UNCLASSIFIED

AD NUMBER

AD460465

NEW LIMITATION CHANGE

TO

Approved for public release, distribution unlimited

FROM

Distribution authorized to U.S. Gov't. agencies and their contractors; Specific Authority; Dec 1964. Other requests shall be referred to Air Force Malterials Lab, WPAFB, OH.

AUTHORITY

afml ltr 6 Dec 1971

THIS PAGE IS UNCLASSIFIED

NOTICE: When government or other drawings, specifications or other data are used for any purpose other than in connection with a definitely related government procurement operation, the U. S. Government thereby incurs no responsibility, nor any obligation whatsoever; and the fact that the Government may have formulated, furnished, or in any way supplied the said drawings, specifications, or other data is not to be regarded by implication or otherwise as in any manner licensing the holder or any other person or corporation, or conveying any rights or permission to manufacture, use or sell any patented invention that may in any way be related thereto.

FUNDAMENTALS OF FORGING PRACTICE

AIR FORCE MATERIALS LABORATORY

BATTELLE MEMORIAL INSTITUTE

A MANUAL

ON

FUNDAMENTALS OF FORGING PRACTICE

By

A. M. SABROFF
F. W BOULGER
H. J. HENNING
J. W. SPRETNAK

SUPPLEMENT TO TECHNICAL DOCUMENTARY REPORT No. ML-TDR-64-95
December 1964

MANUFACTURING TECHNOLOGY DIVISION
AIR FORCE SYSTEMS COMMAND
WRIGHT-PATTERSON AIR FORCE BASE, OHIO

Prepared Under Contract No. AF 33(600)-42963 By
BATTELLE MEMORIAL INSTITUTE, COLUMBUS, OHIO

NOTICES

FOREWORD

This Technical Manual was prepared under United States Air Force Contract No. AF 33(600)-42963. Part of the information used in the compilation and preparation of the Manual was obtained from experimental forging studies conducted as a part of the program and described in Technical Documentary Report No. ML-TDR-64-95, "A Study of Forging Variables".

This contract with Battelle Memorial Institute of Columbus, Ohio, was initiated under Manufacturing Methods Project 7-876, "A Study of Forging Variables". It was administered under the direction of Mr. G. M. Glenn and Mr. G. L. Campbell of the Metallurgical Processing Branch (MATB), Manufacturing Technology Division, Air Force Materials Laboratory, Wright-Patterson Air Force Base, Ohio.

This program was carried out by Battelle's Metalworking Research Division under the supervision of Mr. A. M. Sabroff, Associate Chief, and Mr. F. W. Boulger, Division Chief. Mr. H. J. Henning, formerly of Battelle, conducted the experimental forging studies, compiled the information and data, and assisted in the drafting of the Manual. Dr. J. W. Spretnak, Professor of Metallurgical Engineering at The Ohio State University, served as Consultant to the program, and prepared the section on "Fundamentals of Plastic Deformation" appended to the Manual. The manuscript was released by the authors 31 August 1964 for publication as an RTD Technical Manual.

PREFACE

The technologies of metals and melting processes for metals have advanced rapidly during the past decade. Titanium, beryllium, new nickel-base alloys such as René 41, the precipitation-hardening stainless steels, and the refractory metals all have evolved to commercial status in this time. New melting and alloying techniques had to be developed for many of the new metals, and we now consider commonplace such processes as consumable-electrode arc melting, vacuum induction melting, and vacuum degassing.

Similarly, advances have been made in certain metal-fabricating processes, e.g., impacter forging, explosive forming, and high-velocity forging processes. In general, however, advances in fabricating metals have not been so rapid as the developments in materials and melting processes. This does not mean to say that progress in metal fabrication is actually slow. The forging industry for example has learned in a relatively short time how to make complex parts from titanium, which was once considered practically unforgeable.

However, it is probably true that advances in forging would be more rapid if a more scientific approach were substituted for the trial-and error procedures commonly used in the industry. One very practical method by which a more scientific approach to improving forging technology can be applied is to maintain accurate and detailed records of forging variables. This kind of information can then be used to select forging practices for alloys and parts to be forged in the future. Unfortunately, the recording of detailed forging procedures is expensive, and only a few companies maintain such records. Even these companies have initiated the practice in only relatively recent times.

Recognizing the need for information in this field by engineers, scientists, and technicians in the aerospace and allied industries, the Manufacturing Technology Division of the Air Force Materials Laboratory established a project at Battelle with the objective of preparing this manual on "Fundamentals of Forging Practice". The information presented in the Manual was obtained from the literature, from industrial sources, and from a laboratory-scale experimental program designed to study significant forging characteristics of several typical alloys. This information can be divided into four major subject categories:

(1) Plastic deformation of metals
    This information deals with the mechanics of plastic deformation, and the fundamental principles of metal behavior during deformation.

(2) Principles of forging
    This information concerns the empirical relationships developed for forging processes which serve as practical guides for establishing design limits and shop practices.

(3) Forging processes and practices
    This information relates to the state of the art of forging, and covers such topics as forging equipment, types of operations, die design, lubrications, specifications for typical forged shapes, etc.

(4) Forging alloys
    These data treat the forging of specific alloy systems in terms of the influence of material properties on forging behavior, and the influence of forging procedure on the properties of the forged product.

In organizing this material for the Manual, a major objective was to achieve a document of maximum usefulness to engineers concerned with the design, production, and use of forgings. Accordingly, the main body of the Manual deals with the design, production, and application of die forgings. Chapters 1 through 6 discuss general aspects of the forging process and forging practices; Chapters 7 through 22 then discuss the forging of specific alloy systems. Theories of metal behavior and plastic deformation as they apply to the forging process are discussed in the Appendix for the reader interested in reviewing the fundamental metallurgical and mechanical principles of metal deformation.

A major contributor to the preparation of the Manual was the forging industry. The whole-hearted support of many companies in this industry has made it possible to present up-to-date information on a number of comparatively new materials, and has added immeasurably to the value of the Manual. It also has indicated the pressing need for compiling this information in the form of technical manuals such as this.

AMS/FWB/HJH/JWS

ACKNOWLEDGMENT

This Manual was made possible by the support and cooperation of the forging and aerospace industries, to whom the authors are indebted for their willingness to provide unpublished data and information. A special note of thanks is due the Wyman-Gordon Company, Worcester, Massachusetts, and the Ladish Company, Cudahy, Wisconsin, for their cooperation and assistance in reviewing the early drafts of sections of the Manual. The authors would also like to thank the many members of the Process and Physical Metallurgy Department at Battelle who assisted in reviewing the final draft of the Manual.

TABLE OF CONTENTS

	Page
FOREWORD	
PREFACE	
CHAPTER 1. FORGING PROCESSES AND EQUIPMENT	1
Types of Forging Operations	1
Types of Forging Equipment	1
Drop and Counterblow Hammers	4
Hydraulic Presses	4
Mechanical Presses	12
Impacters	13
High-Energy-Rate Forging Machines	13
Ring-Rolling Mills	18
Swaging Machines	18
Rotary Forging Machines	18
Shear-Forming Machines	21
Choice of Hammers or Presses	24
References	26
CHAPTER 2. METAL FLOW IN FORGING	27
Development of Grain Flow	27
Control of Grain Flow	30
Metal Flow in Impression Dies	33
Effects of Nonuniform Deformation	33
Estimating the Amount of Deformation	37
Factors Influencing Die Filling	43
Forgeability and Flow Strength	43
Friction and Lubrication	46
Die Temperature	50
Shape and Size Factors	53
References	58
CHAPTER 3. PRINCIPLES OF DIE-FORGING DESIGN	59
Design Factors	59
Parting-Line Location	59
Finish Allowance	61
Forging Draft	62
Flash	64
Fillet and Corner Radii	66
Minimum Section Size	69
Forging Tolerances	72
References	77
CHAPTER 4. FORGING-DESIGN PRACTICES	79
Establishing the Initial Forging Design	79
Designs for Disks, Cones, and Other Circular Shapes	81
Designs for Structural Shapes	87
Designs for Forgings With Precision Tolerances	99
Precision Forging	99
Target Machining	110
References	113
CHAPTER 5. FORGING DEFECTS	115
Classification of Defects	115
Laps	115
Coarse-Grain Wrinkles	116
Flow-Through Defects	116
Hot Tears and Center Bursts	117

TABLE OF CONTENTS
(Continued)

	Page
Disturbed-Metal Defects	117
Thermal Cracks	117
Inspection Methods	119
CHAPTER 6. FORGEABILITY OF METALS AND ALLOYS	121
Definition of Forgeability	121
Measurement of Forgeability	121
Metallurgical Factors Influencing Forgeability	127
Mechanical Factors Influencing Forgeability	132
Methods of Improving Forgeability	136
Predicting Forging Behavior	140
References	142
CHAPTER 7. ALUMINUM FORGING ALLOYS	145
Alloys and Forms Available	145
Forging Behavior	145
Influence of Forging Variables on Mechanical Properties	148
Commentary on Forging Practices	151
Commentary on Forging Design	153
References	154
CHAPTER 8. MAGNESIUM FORGING ALLOYS	155
Alloys and Forms Available	155
Forging Behavior	157
Influence of Forging Variables on Mechanical Properties	158
Commentary on Forging Practices	160
Commentary on Forging Design	161
References	162
CHAPTER 9. CARBON AND LOW-ALLOY STEELS	165
Alloys and Forms Available	165
Forging Behavior	165
Forgeability	165
Flow Stress and Forging Pressure	166
Influence of Forging Variables on Mechanical Properties	172
Commentary on Forging Practices	174
Commentary on Forging Design	177
References	178
CHAPTER 10. MARTENSITIC STAINLESS STEELS	179
Alloys and Forms Available	179
Forging Behavior	180
Commentary on Forging Practices and Design	182
References	184
CHAPTER 11. AUSTENITIC STAINLESS STEELS	185
Alloys and Forms Available	185
Forging Behavior	186
Commentary on Forging Practices	190
References	192
CHAPTER 12. PRECIPITATION-HARDENABLE STAINLESS STEELS	193
Alloys and Forms Available	193
Forging Behavior	195
Influence of Forging Variables on Mechanical Properties	196
Commentary on Forging Practices	197
References	201

TABLE OF CONTENTS
(Continued)

	Page
CHAPTER 13. MARAGING STEELS	203
Alloys and Forms Available	203
Forging Behavior	204
Influence of Forging Variables on Mechanical Properties	205
Commentary on Forging Practice	207
References	209
CHAPTER 14. TITANIUM AND TITANIUM ALLOYS	211
Alloys and Forms Available	211
Forging Behavior	212
Influence of Forging Variables on Mechanical Properties	215
Commentary on Forging Practices	228
Commentary on Forging Design	230
References	232
CHAPTER 15. IRON-BASE SUPERALLOYS	235
Alloys and Forms Available	235
Forging Behavior	236
Commentary on Forging Practices	238
References	244
CHAPTER 16. COBALT-BASE SUPERALLOYS	247
Alloys and Forms Available	247
Forging Behavior	248
Commentary on Forging Practices	249
References	251
CHAPTER 17. NICKEL-BASE SUPERALLOYS	253
Alloys and Forms Available	253
Forging Behavior	257
Influence of Forging Variables on Mechanical Properties	264
Commentary on Forging Practices	268
Commentary on Forging Design	270
References	275
CHAPTER 18. COLUMBIUM AND COLUMBIUM ALLOYS	277
Alloys and Forms Available	277
Forging Behavior	278
Commentary on Forging Practices	282
References	286
CHAPTER 19. TANTALUM AND TANTALUM ALLOYS	287
Alloys and Forms Available	287
Forging Behavior	288
Commentary on Forging Practices	292
References	293
CHAPTER 20. MOLYBDENUM AND MOLYBDENUM ALLOYS	295
Alloys and Forms Available	295
Forging Behavior	295
Influence of Forging Variables on Mechanical Properties	303
Commentary on Forging Practices	309
Commentary on Forging Design	[illegible]
References	314
CHAPTER 21. TUNGSTEN AND TUNGSTEN ALLOYS	317
Alloys and Forms Available	317
Forging Behavior	318
Commentary on Forging Practices	322

TABLE OF CONTENTS
(Continued)

	Page
Commentary on Forging Design	327
References	332
CHAPTER 22. BERYLLIUM	335
Grades and Forms Available	335
Forging Behavior	336
Commentary on Forging Practices	342
Canned-Powder Forging	342
Billet Forging	345
Commentary on Forging Design	351
References	354
APPENDIX A	
FUNDAMENTALS OF PLASTIC DEFORMATION	A-1
Mechanical Response of a Perfect Elastic Solid	A-2
Theoretical Shear Strength	A-2
Theoretical Fracture Strength	A-3
The Rheological Models for Real Solids	A-5
Defect Structure in Real Metals	A-8
Point Defects	A-8
Line Defects	A-10
Surface Defects	A-10
Elements of Dislocation Theory	A-12
Deformation Processes in Metals	A-15
Deformation Mechanisms in the Low-Temperature Range	A-15
Deformation Mechanisms in the High-Temperature Range	A-20
Macroscopic Deformation	A-21
Deformation of Single Crystals	A-21
Deformation of Polycrystals	A-26
Tensile Deformation	A-29
Variables Influencing the Flow Curve	A-31
Equivalence of Strain Rate and Temperature	A-34
Simplified Models of Flow and Fracture	A-35
Fracture of Single Crystals	A-37
Fracture of Polycrystalline Metals	A-38
Fracture Nomenclature and Mechanisms	A-38
Relationship Between Elastic and Plastic Properties of Pure Polycrystalline Metals	A-40
Cold Working and Hot Working	A-43
Cold Working	A-43
Recrystallization	A-46
Hot Working	A-47
Work of Deformation	A-48
Generation of Heat	A-51
Hardening Mechanisms and Other Metallurgical Effects	A-52
Hardening by Alloying	A-52
Stability of Microstructures at Elevated Temperatures	A-56
Oxidation and Contamination	A-57
Embrittlement of Metals and Alloys	A-58
Residual Stress	A-59
Anisotropy in Wrought Alloys	A-60
REFERENCES – APPENDIX A	A-61
APPENDIX B	
SELECTED BIBLIOGRAPHY ON FORGING PRACTICES	B-1

TABLE OF CONTENTS
(Continued)

Page

APPENDIX C

SELECTED BIBLIOGRAPHY ON PLASTIC DEFORMATION AND METAL FLOW THEORY. C-1

APPENDIX D

SELECTED BIBLIOGRAPHY ON ELEVATED-TEMPERATURE PROPERTIES OF METALS AND ALLOYS. D-1

APPENDIX E

SELECTED BIBLIOGRAPHY ON FORGING LUBRICANTS E-1

CHAPTER 1

FORGING PROCESSES AND EQUIPMENT

TYPES OF FORGING OPERATIONS

While forging may be defined broadly as changing the shape of metals by plastic deformation, there are a number of specific types of forging operations that, in many cases, require specialized equipment. Descriptions of several types of forging operations and the specialized equipment needed for them are presented in Table 1-1, and the principles are illustrated in Figure 1-1.

The forces required to deform metal in each of these operations differ considerably, depending on the relative amount of confinement of the workpiece by the dies. With increasing confinement, friction increases rapidly. Furthermore, when workpiece temperatures are higher than that of the dies, heat transfer occurs and billet surfaces are chilled. Both factors increase forging pressures. During the upsetting, drawing-out, swaging, and ring-rolling operations, there is considerable latitude for metal to flow freely since there is little confinement. On the other hand, during core forging, die forging, and extrusion operations, the metal is so confined by the dies that there is little or no free metal flow. Extrusion forging, for example, approaches complete confinement at the end of the forging stroke. This results in considerably greater resistance to metal flow. Thus, a metal which can be deformed easily when frictional loads are small requires relatively high forging pressures during extrusion forging.

TYPES OF FORGING EQUIPMENT

In a book published in 1938, Naujoks and Fabel described forging-plant equipment in considerable detail.(1)* They covered forging machinery, shearing equipment, die-sinking equipment, furnaces, and other auxiliary process equipment found in forging plants at that time. Since then, several new types and new modifications of forging equipment have been introduced. Hydraulic presses have increased in size, the counterblow hammer was introduced in this country, The Chambersburg Cecomatic Impacter was developed and, more recently, high-velocity, pneumatic-mechanical machines have been used for forging. Larger and more versatile mechanical presses have also increased the capability of the forging industry.

The behavior of metals during forging is influenced by the time necessary to complete the plastic shaping. Thus, it is important to recognize that the basic difference between the types of equipment lies in their forging velocities, or rates of deformation (not to be confused with cycle rates). Forging hammers, for instance, deform metals at rates of deformation on the order of 100 times faster than do hydraulic presses. The following sections describe these two major types of forging equipment, and other, more specialized forging machinery.

*References are listed at the end of this chapter.

TABLE 1-1. DESCRIPTION OF AND MACHINERY USED FOR SEVERAL COMMON TYPES OF FORGING OPERATIONS

Type of Forging	Method of Operation	Commonly Used Machinery
Upsetting	Compression parallel with the longitudinal axis of the work	Single-action and counterblow hammers Upsetting machines Hydraulic, air, and mechanical presses High-energy-rate machines
Drawing out	Stretching of the work by a series of upsets along the length of the workpiece	Single-action hammers Hydraulic and air presses
Die forging	Compression in a closed, impression die	Single-action and counterblow hammers Hydraulic and mechanical presses High-energy-rate machines Impacters
Ring rolling	Radial compression on a ring shape to increase diameter	Ring-rolling mills Hammers and presses with supported mandrel
Swaging	Radial compression by shaped dies to lengthen a workpiece	Swaging machine Single-action hammers Air and hydraulic presses
Core forging	Displacing metal with a punch to fill a die cavity	Multiple-ram presses
Extrusion forging	Forcing metal into a die opening by restricting flow in other directions	Hydraulic and mechanical presses Multiple-ram presses High-energy-rate machines
Back extrusion	Forcing metal to flow in a direction opposite to the motion of the punch with respect to the die	Single-action and counterblow hammers Hydraulic and mechanical presses Multiple-ram presses High-energy-rate machines

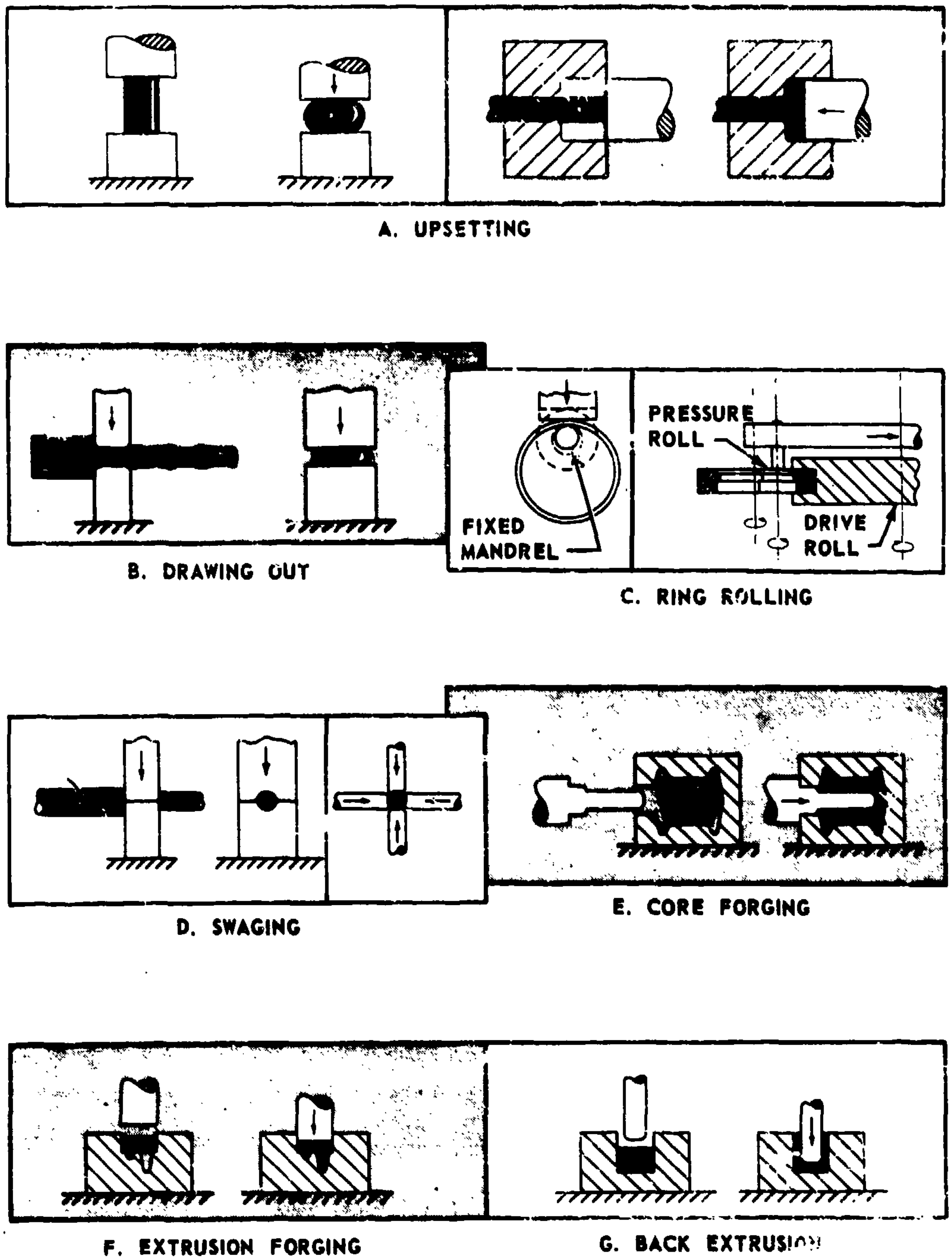


FIGURE 1.1. PRINCIPLES OF SEVERAL TYPES OF FORGING OPERATIONS

Drop and Counterblow Hammers

For drop hammers the necessary force and energy are supplied by a falling weight or ram, usually assisted by pressurized air or steam. Production drop hammers range in size upwards from about 500 pounds. The largest known single-action drop hammer is a 50,000-pound steam hammer. Ram velocities of drop hammers are usually on the order of 150 to 220 inches per second, depending on stroke length and whether or not the basic gravity power is assisted by steam or air. During forging, the ram strikes the workpiece with repeated blows and thus shapes the metal in a stepwise fashion. Drop hammers are equipped with an anvil or static base 10 to 20 times heavier than the ram.

The operation of counterblow hammers is quite similar to that of drop hammers except that two rams are activated simultaneously in opposite directions. They strike repeated blows at a midway point and develop combined velocities about 1-1/2 the normal hammer velocities. The equalizing dynamic forces of the opposed moving rams eliminate the need for the heavy bases characteristic of single-action drop hammers. The largest counterblow hammer in this country is said to develop energy equivalent to that of a 150,000-pound drop hammer. Examples of both types of forging hammers are shown in Figures 1-2 and 1-3.

The ratings of gravity drop hammers are based on the energy developed when the falling ram strikes a workpiece at the velocity determined by the height of the fall. To develop higher energy, pressurised steam or air is used to drive the ram at higher velocities. Thus, shorter strokes can be used to achieve a given energy, or more energy can be delivered for a given stroke by increasing the ram velocity. Hence, larger hammers are usually of the steam type.

The ratings of the counterblow hammers are based on ram weight and the combined velocity of the opposed steam-driven rams. Since the combined velocities achievable are greater than those obtainable in single-action hammers, ratings based only on ram weight are not directly comparable. Essentially, the counterblow principle is another means of increasing the energy available for hammer forging without necessarily changing ram weight.

Hydraulic Presses

Hydraulic presses are operated by large pistons driven by high-pressure hydraulic or hydropneumatic systems. They are usually slow moving (up to 2 inches per second) under pressure. The two largest presses in this country are rated at 50,000 tons: one is at the Wyman-Gordon Company, North Grafton, Massachusetts, and the other is at the Aluminum Company of America, Cleveland, Ohio. Smaller production presses range in capacity between about 300 tons and 35,000 tons. Typical hydraulic forging presses are shown in Figures 1-4 through 1-8.

Among the wide variety of hydraulic presses available in the forging industry, there are two basic types:

(1) Direct-drive hydraulic presses. These presses operate with hydraulic fluid (oil or water) pressurized directly by high-pressure pumps. Larger presses require proportionally larger pumps. Hence, the capacity of these presses is limited to some extent by practical limits on pump capacities.

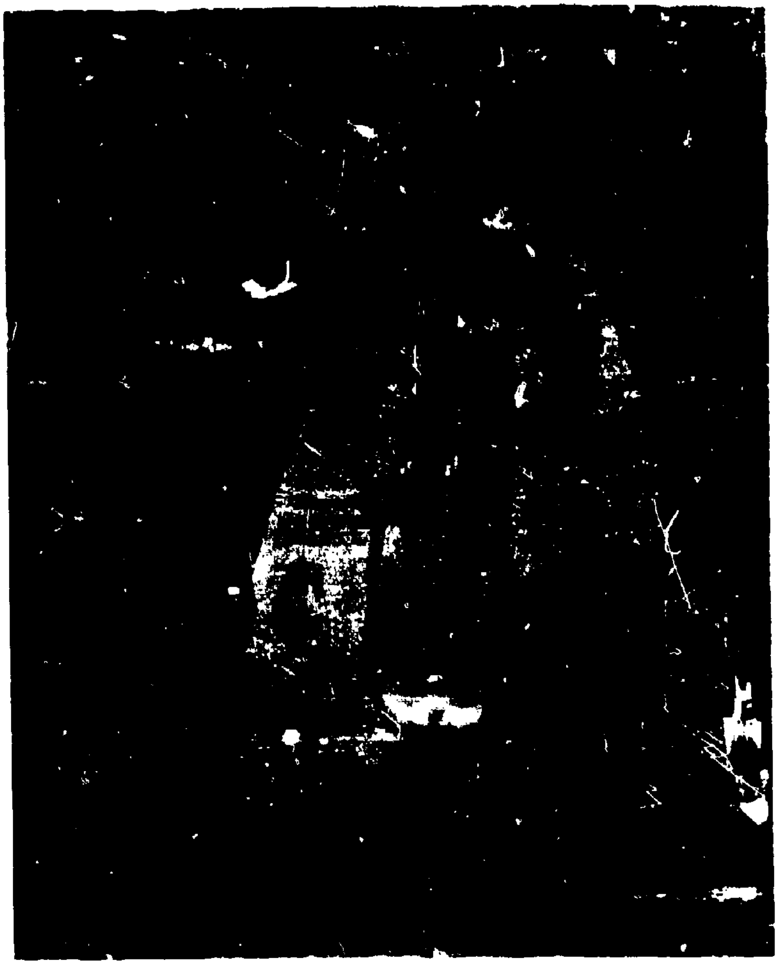


FIGURE 1-2. A 20,000-POUND STEAM DROP HAMMER

Courtesy of Wyman-Gordon Company.

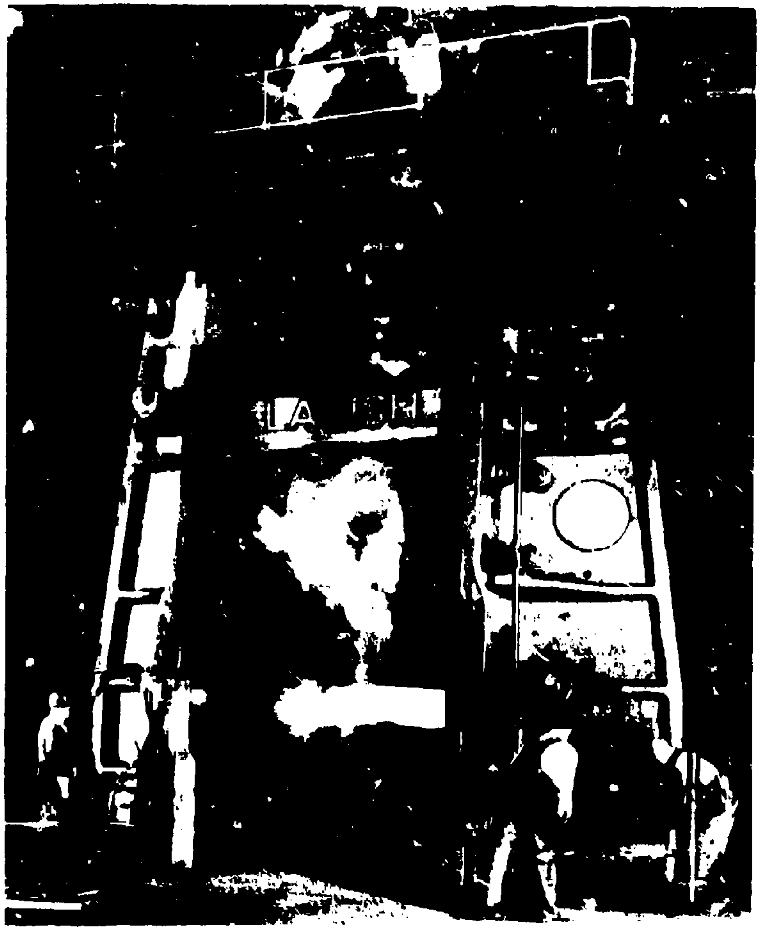


FIGURE 1-3. A LARGE COUNTERBLOW HAMMER WITH A RATING EQUIVALENT TO ABOUT A 150,000-POUND DROP HAMMER

Courtesy of Ladish Company.

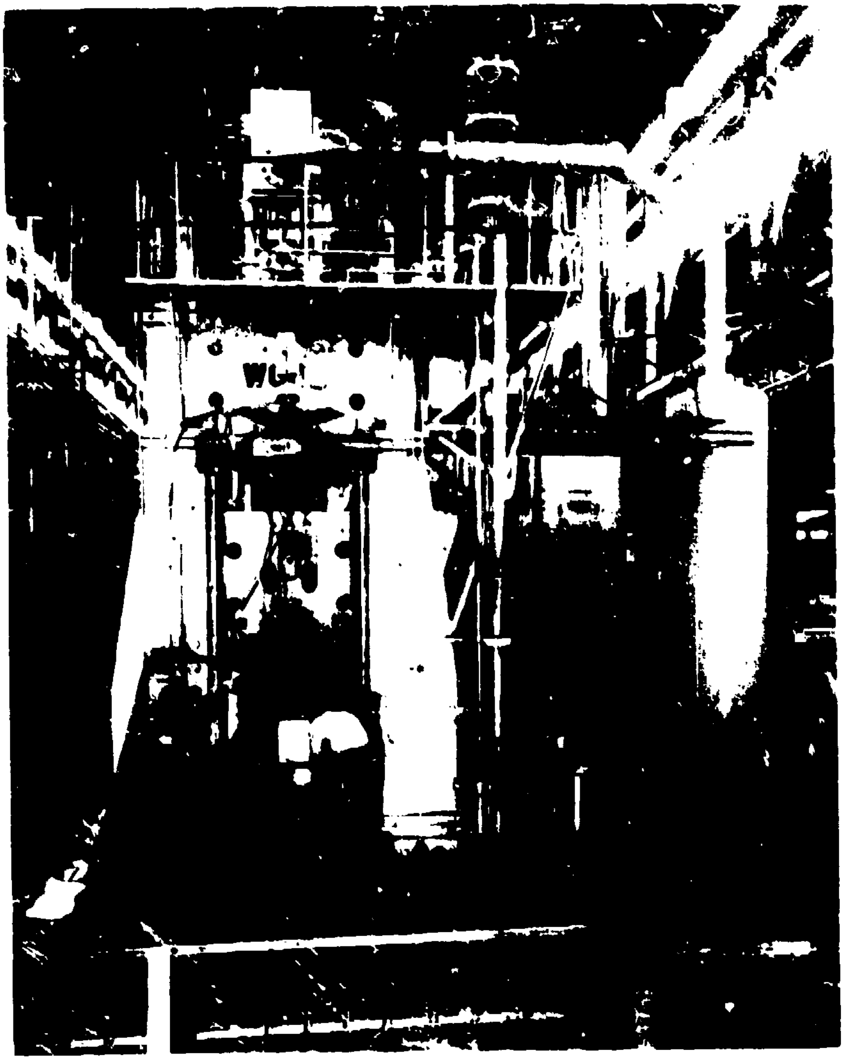


FIGURE 1-4. 1500-TON SINGLE-RAM HYDRAULIC FORGING PRESS

Courtesy of Wyman-Gordon Company.

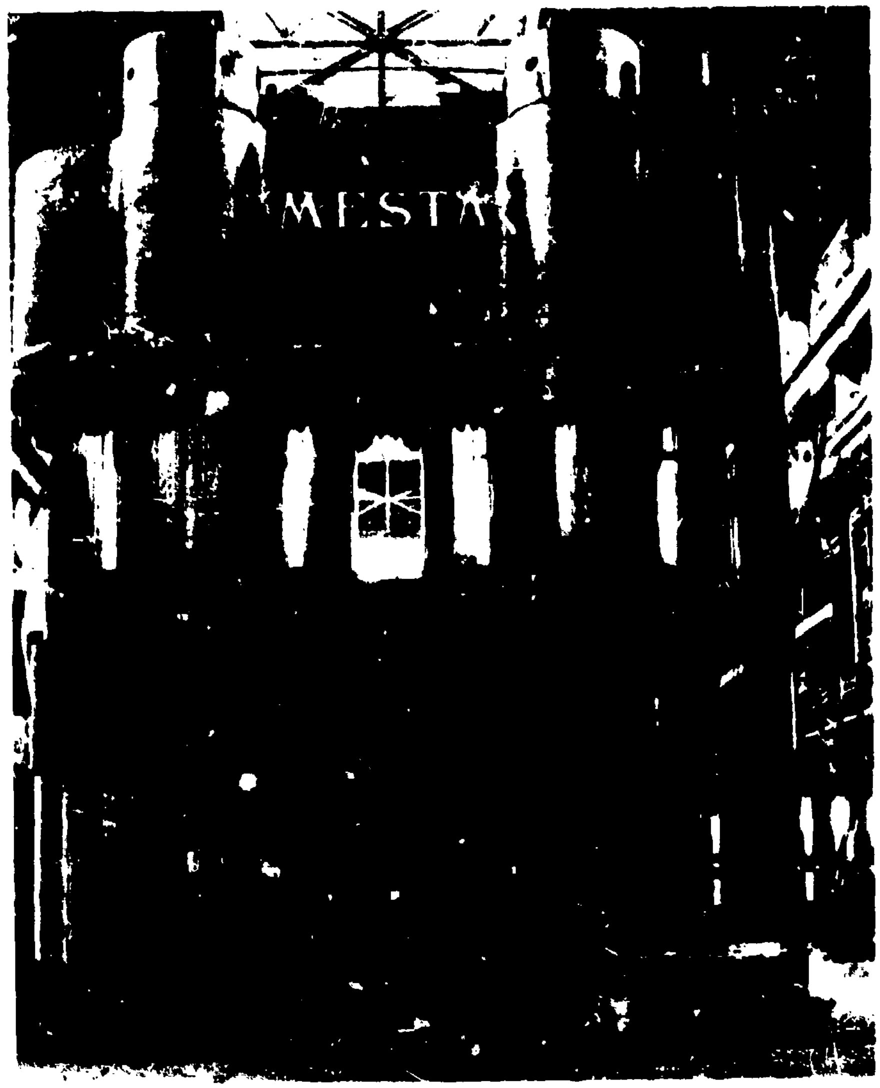


FIGURE 1-5. 18,000-TON SINGLE-RAM HYDRAULIC FORGING PRESS

Courtesy of Wyman-Gordon Company.

FIGURE 1-6. MODERN 20,000-TON HYDRAULIC FORGING PRESS

Courtesy of Cameron Iron Works.

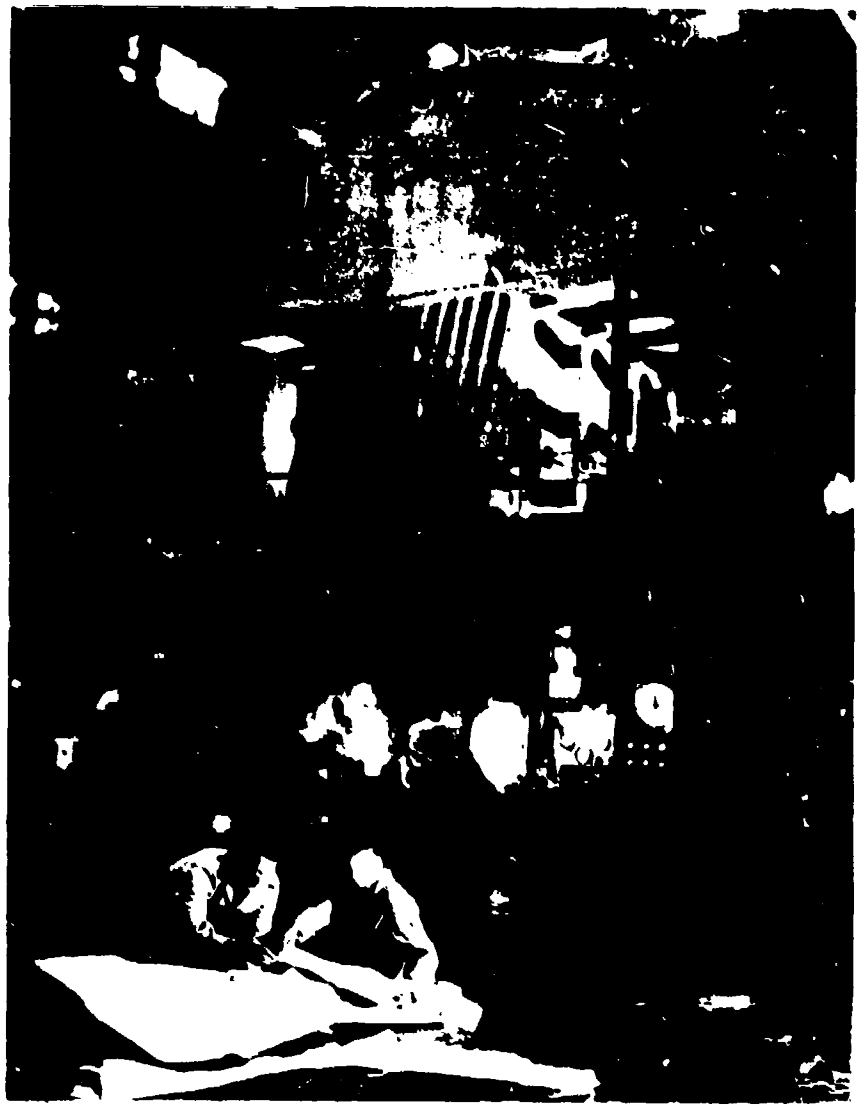


FIGURE 1-7. 50,000-TON HYDRAULIC FORGING PRES

Courtesy of Wyman-Gordon Company.

FIGURE 1-6. 11,000-TON MULTIPLE-RAM HYDRAULIC FORGING PRESS

Courtesy of Cameron Iron Works.

(2) Hydropneumatic presses. These presses operate with hydraulic fluid supplied from accumulators which are in turn pressurized by high-pressure pumps. The larger presses are of this type.

A slight disadvantage of the hydropneumatic system is that the accumulator pressure drops below maximum capacity as the press ram advances. Direct-drive hydraulic presses maintain maximum pressure throughout the pressing cycle. Both types of hydraulic press require several seconds, or more, to complete a forging cycle. When forging metals heated to temperatures well above that of the dies, a comparatively long forging cycle can cause chilling of the workpiece and/or overheating of the dies.

The multiple-ram forging press offers advantages not available in single-action press-forging equipment or in drop-hammer equipment. A typical multiple-ram hydraulic forging press, shown in Figure 1-8, can be used to forge hollow parts in one pressing. Figure 1-9 shows the successive steps in forging a double-cone venturi nozzle, demonstrating how the main ram serves to hold two split dies together while the side rams force heated metal to take the shape of a part impossible to forge in a single-action press.

(a) Heated Billet in Position on Die for Forging

(b) Main Ram and Side Rams in Final Position During Forging

(c) Finished Forging

FIGURE 1-9. MULTIPLE-RAM PRESS FORGING OF A VENTURI NOZZLE SHAPE

Courtesy of Cameron Iron Works.

Mechanical Presses

Mechanical presses differ from hammers and hydraulic presses in that they force two working surfaces together by offset cams, cranks, and other rigidly connected mechanical systems. The strokes of mechanical presses are shorter than those of either hammers or hydraulic presses. For this reason, mechanical presses are often favored for low-profile forgings. Mechanical presses range in size upwards to nearly 10,000-ton capacity.

An advantage of mechanical presses is that they have accurately controlled strokes; therefore, closer tolerances can usually be expected. The greatest forging loads are developed at the end of the stroke. A disadvantage, compared with hammers, is that the forging action is slow, ram speeds being in the vicinity of 1 to 10 in./sec. For this reason, die-chilling effects impose restrictions on forging designs in much the same manner as for hydraulic presses.

Mechanical presses are commonly used in automated forge shops where large production quantities are made with each tooling setup. Their greatest use in job shops is for trimming forgings made on the more versatile hammers and presses.

Impacters

Impacters consist of opposed hammers that operate on the same principle as the counterblow hammer except that the rams operate horizontally.[2] This equipment is designed specifically for producing large quantities of low-profile forgings. The rams are said to strike at speeds about equal to those of hammers, with energy ranging upward to the equivalent of a 5000-pound drop hammer. Equipment of this kind is found in specialty forge shops* that produce large volumes of a particular type of part. Impacters are not often used by commercial forge shops.

Like mechanical presses, impacters are readily adapted to automation. Because the faster ram velocities minimize die chilling, impacters are useful for producing forgings with thinner sections. Heavy hammer bases are not needed for impacters because the opposed moving rams react with each other during the forging process.

During forging, the part is held between the dies by a manipulator set to position the forging, automatically, for successive blows. For this reason, forgings designed for impacters require an added piece of metal for handling purposes.

High-Energy-Rate Forging Machines

High-energy-rate forging machines are based on essentially the same principles as are drop hammers and impacters. They provide greater energies for a given ram weight by using ram velocities on the order of 2 to 10 times those of hammers. One machine with a 2000-pound ram (Model 210 Dynapak) is said to develop a striking energy comparable with that of a 10,000-pound drop hammer. One of the most significant advantages of these machines is that the ram velocity (hence the impacting energy) is highly reproducible. Forgings may be made to close tolerances without the die-to-die contact necessary for dimensional control in hammer forging. Because die-contact times are short, parts with unusually thin sections may be forged in these machines. As a result of the high strain rates, the metal temperatures may actually increase significantly, depending on the amount of deformation that takes place. This actually becomes a limiting factor when forging metals with narrow forging-temperature ranges.

High-energy-rate forging machines all operate on essentially the same principle: the sudden release of a stored, compressed gas against a free piston. However, the methods of releasing the gas to actuate the piston and, accordingly, the machine construction have changed as improvements and innovations have been incorporated in the basic designs.

*Distinction is made here between specialty and commercial forge shops. Specialty forge shops usually limit their business to a few types of parts while the commercial forge shops handle any number of shapes and sizes of forgings. Captive forge shops such as those in the automobile industry are examples of specialty shops since they produce such parts, e.g. connecting rods, crankshafts, gears, etc., on a continuous basis.

All of the machines are self-reacting, that is, the momentum of the ram is offset by an equal and opposite momentum of the anvil or platen. The Dynapak machine is similar in this respect to drop hammers in that the energy of a fast-moving, lightweight ram is reacted by a slow-moving, heavier platen which is also the machine frame. The Hermes machine, on the other hand, is more like an impacter in that the energy of the upper ram is balanced by an opposing ram of equal weight moving at the same velocity. Because of this self-reacting principle, the machines do not require massive foundations as do conventional drop hammers.

Early models of the Dynapak machine, as shown in Figure 1-10, utilize tie-rod construction for the frame. The main cylinder consists of two chambers: one for storing the compressed gas and the other for housing the piston assembly. The piston is raised to the "firing" position by hydraulic fluid which is then replaced by low-pressure gas. The piston is then "fired" by opening a quick-release valve in the upper chamber and exposing the piston face to the high-pressure gas. The machine assembly is attached to the housing through shock absorbers mounted on the tie rods. Thus, as the ram advances, the heavier machine assembly moves in the opposite direction at a slower speed, but with equal momentum.

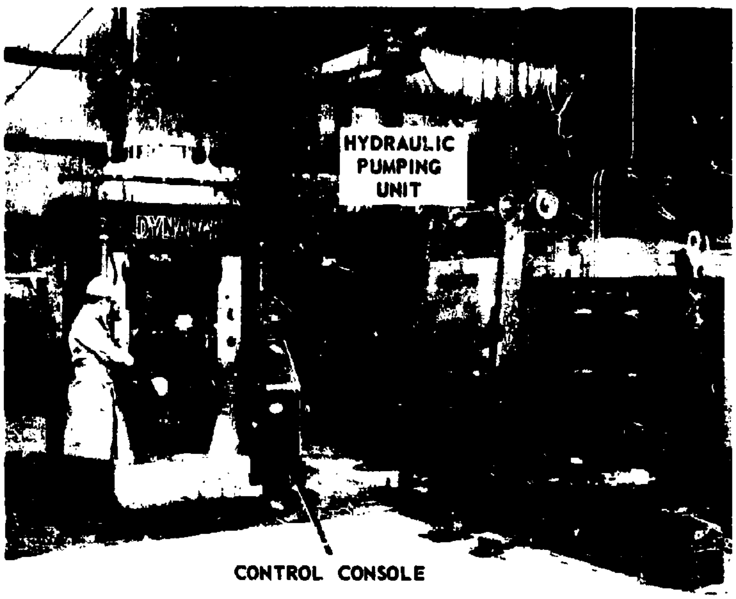


FIGURE 1-10. TYPICAL HIGH-ENERGY-RATE MACHINE INSTALLATION

Courtesy of General Dynamics.

Later models of the Dynapak are built with a yoke type of frame construction, as shown in Figure 1-11. With this design, a single chamber houses both the compressed gas and the piston. The piston is "cocked" and sealed against the top of the chamber by two jacks mounted in the base of the frame. With high-pressure gas in the chamber, the

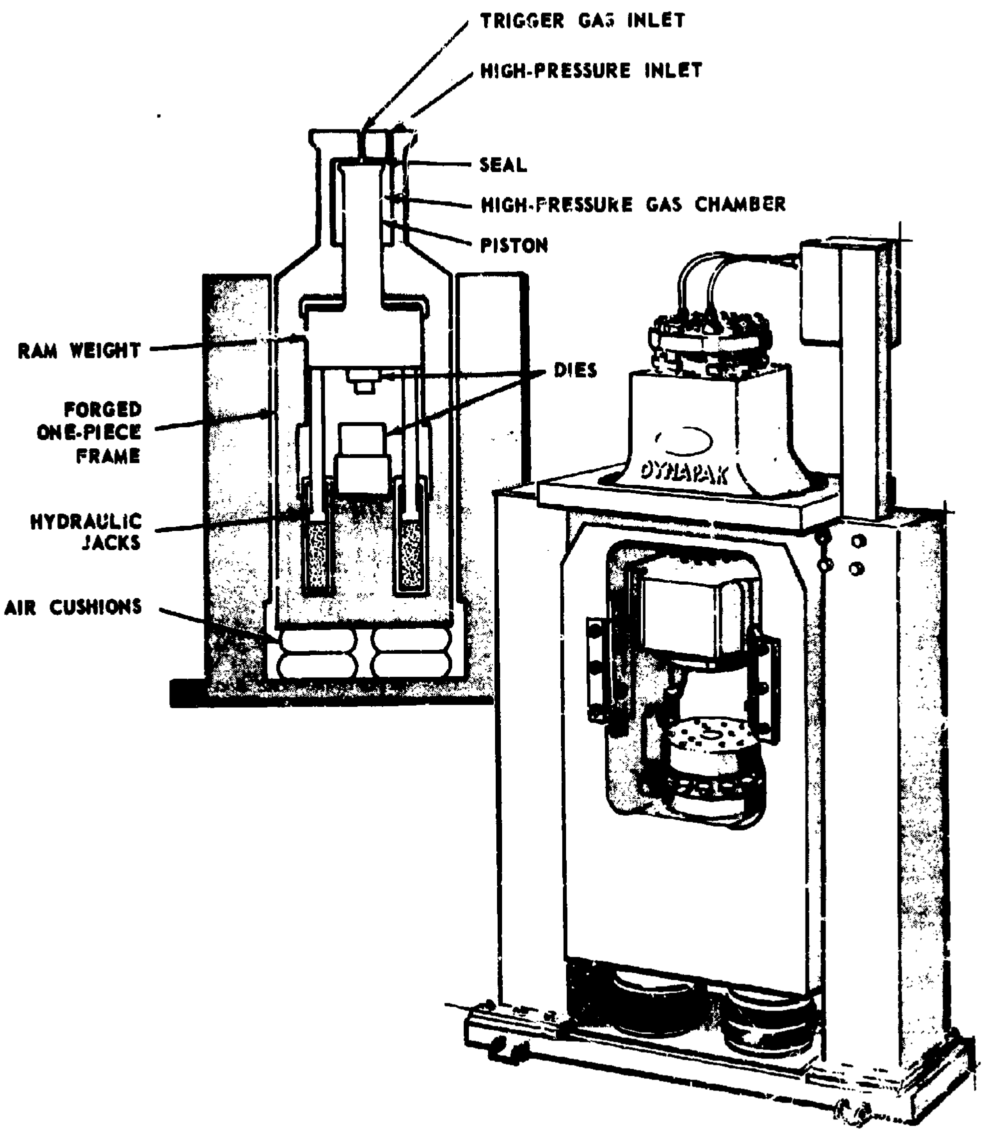


FIGURE 1-11. SCHEMATIC DIAGRAM OF DYNAPAK MODEL 1220 C HIGH-ENERGY-RATE FORGING MACHINE SHOWING ESSENTIAL COMPONENTS

COURTESY ADVANCED PRODUCTS DEPARTMENT OF GENERAL DYNAMICS CORPORATION

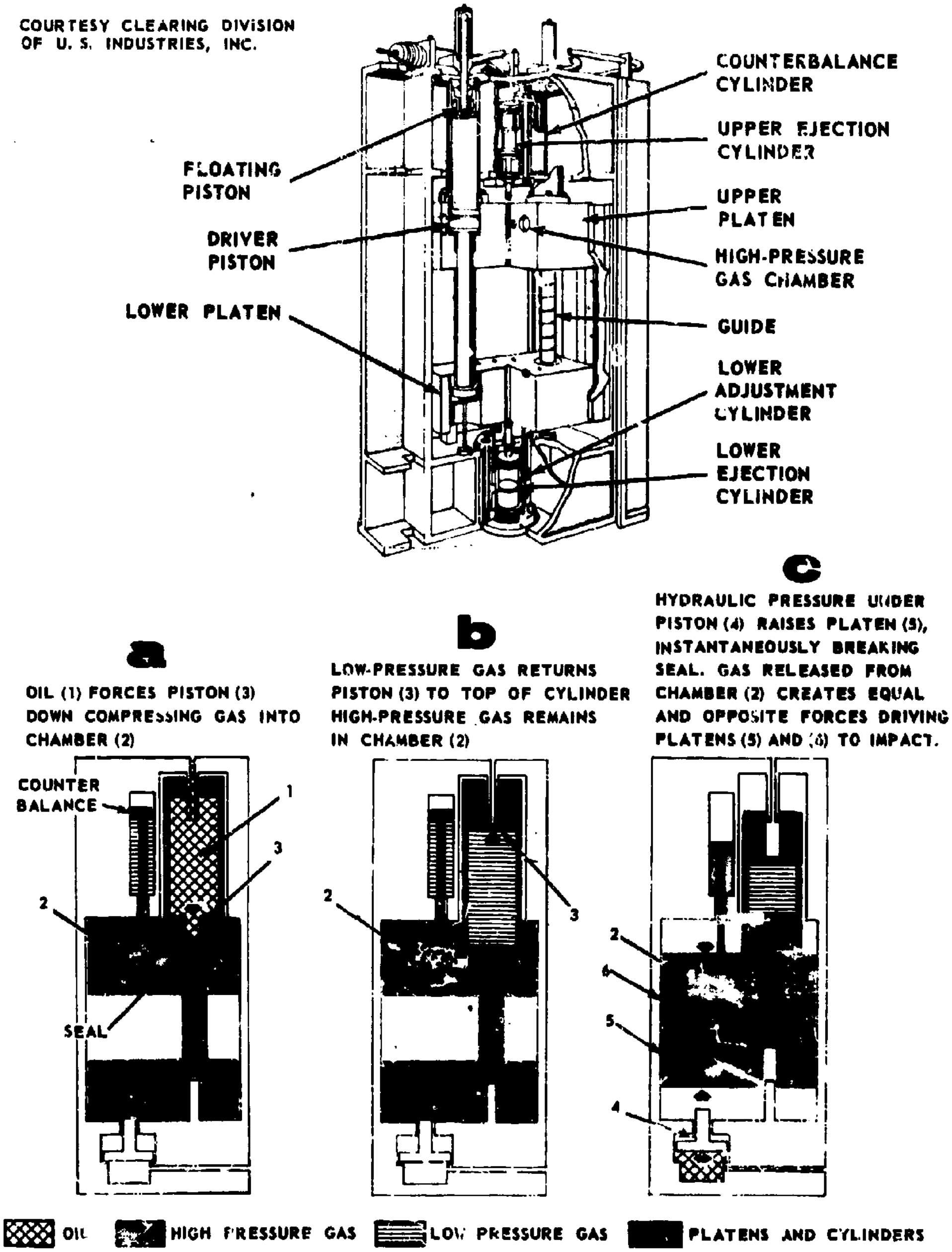


FIGURE 1-12. CUTAWAY VIEW OF HERMES MODEL 2500 B HIGH-ENERGY-RATE MACHINE

ram is held in position by the gas pressure and the machine is in static balance. The piston seal is then broken by injecting gas between the top of the chamber and the piston, thereby exposing the piston face to the high-pressure gas and driving it downward. At the same time, the frame is moved upward. Air cushions are used to support the floating frame and return it to its original position at the end of the stroke.

In the Hermes machine, shown in cutaway view in Figure 1-12, high-pressure gas is stored in a chamber in the upper platen. As the gas is released, it acts upon the drive piston attached to the lower platen and upon the upper platen at the same time. This force drives the two opposing platens toward each other at the same speed. As the platens close, gas is compressed in a separate cylinder containing a piston connected to the upper platen. The high-pressure gas in this counterbalancing cylinder then returns the platens to their original position.

Compressed nitrogen gas is used for these machines because it is noncorrosive, readily available, and will not create a fire hazard when mixed with oil vapor. Very small amounts of nitrogen are consumed in each cycle, which is not an appreciable cost factor.

The velocity and energy of impact of high-energy-rate machines are controlled by the gas pressure and the mass of the ram assembly. Machines of this type mounted either vertically or horizontally are built with strokes up to 10 inches, capacities up to about 250,000 ft-lb, and ram velocities up to about 100 ft/sec.

Because these machines utilize a high velocity in place of a large mass to develop energy, there is a considerable saving in weight over conventional hammers and presses. For example, the weight of a moving ram can be reduced 95 per cent by a fivefold increase in velocity to obtain the same energy level.

Forging dies for these machines need not be as massive as for drop-hammer forging but design is more critical because of the impact loading. Die materials must be selected for maximum toughness and shock resistance. Properly designed tools will not contact one another during the forging stroke. Rather, the design should permit all energy to be absorbed by the workpiece. This practice avoids die failure. It is important to consider the following points when forging in high-energy-rate machines:

(1) The machine must not be fired unless a workpiece is placed between the dies.

(2) For forgings that require restriking, it is important to reduce the energy, by lowering gas pressure, enough to do just the work required. Otherwise, die failures are likely.

(3) The centers of gravity of tools and workpiece must be aligned axially. This avoids side thrust that may damage the machines.

(4) Dies should be designed so that the forging either can be quickly removed from the dies, or can repeatedly stay in one of them. This practice increases die life and permits close tolerances in at least one die.

(5) Forgings may be made using multiple blows with essentially the same techniques and design principles as used in hammer forging.

(6) Energies available for forging are readily approximated from the familiar formulas for kinetic energy ($E = mv^2$) and isothermal gas expansion ($p_1v_1 = p_2v_2$).

Tests conducted at Battelle indicated that the Model 1210 Dynapak attains actual velocities approximately 80 per cent of those calculated over a fire pressure range of 400 psig to 1200 psig. It is assumed that most machines would perform similarly.

It should be recognized that improvements are continually being made on high-energy-rate machines to improve the stroke-cycle rate. Most of these improvements, however, do not alter the basic principle of operation, only the means for accelerating and returning the ram.

Ring-Rolling Mills

Ring-rolling mills operate on the same principle as horizontal bar-rolling mills except that the rolling operation starts with a thick-wall ring-shaped blank instead of a billet. By continually reducing the wall thickness, the circumference of the ring is enlarged. Forging companies having these mills will usually die forge the ring-shaped blanks in presses or hammers. As indicated in Figure 1-13, rings are frequently rolled to contours similar to those of channels, I-beams, angles, and other shapes producible in conventional rolling mills. The principal advantage lies in the fact that the circular shapes need not be welded.

Rolled rings are also used to produce blanks for the shear-forming process. Some shapes and designs producible by ring rolling are illustrated in Chapter 4.

Swaging Machines

The most common type of swaging machine consists of a power-driven ring that revolves at high speed causing rollers to engage cam surfaces and force the segmented dies to deliver hammer-like blows on the work at high frequencies. The machines are useful for reducing rods without resorting to expensive rolling operations. They can produce both straight and tapered cylindrical sections.

Rotary Forging Machines

Rotary forging machines operate on a principle similar to that of swaging machines, but they are more versatile. They can be used to work bar and tubing into a variety of round shapes having varying diameters and tapers along the axis. These machines are used frequently to forge short runs of shapes that otherwise would be extruded or upset forged, the principal advantage being that tooling is less expensive than for either of the latter operations.[3]

The machines consist of a base which holds the hammers, supporting framework, and a rotary spindle which holds the workpiece (see Figure 1-14). The workpiece is

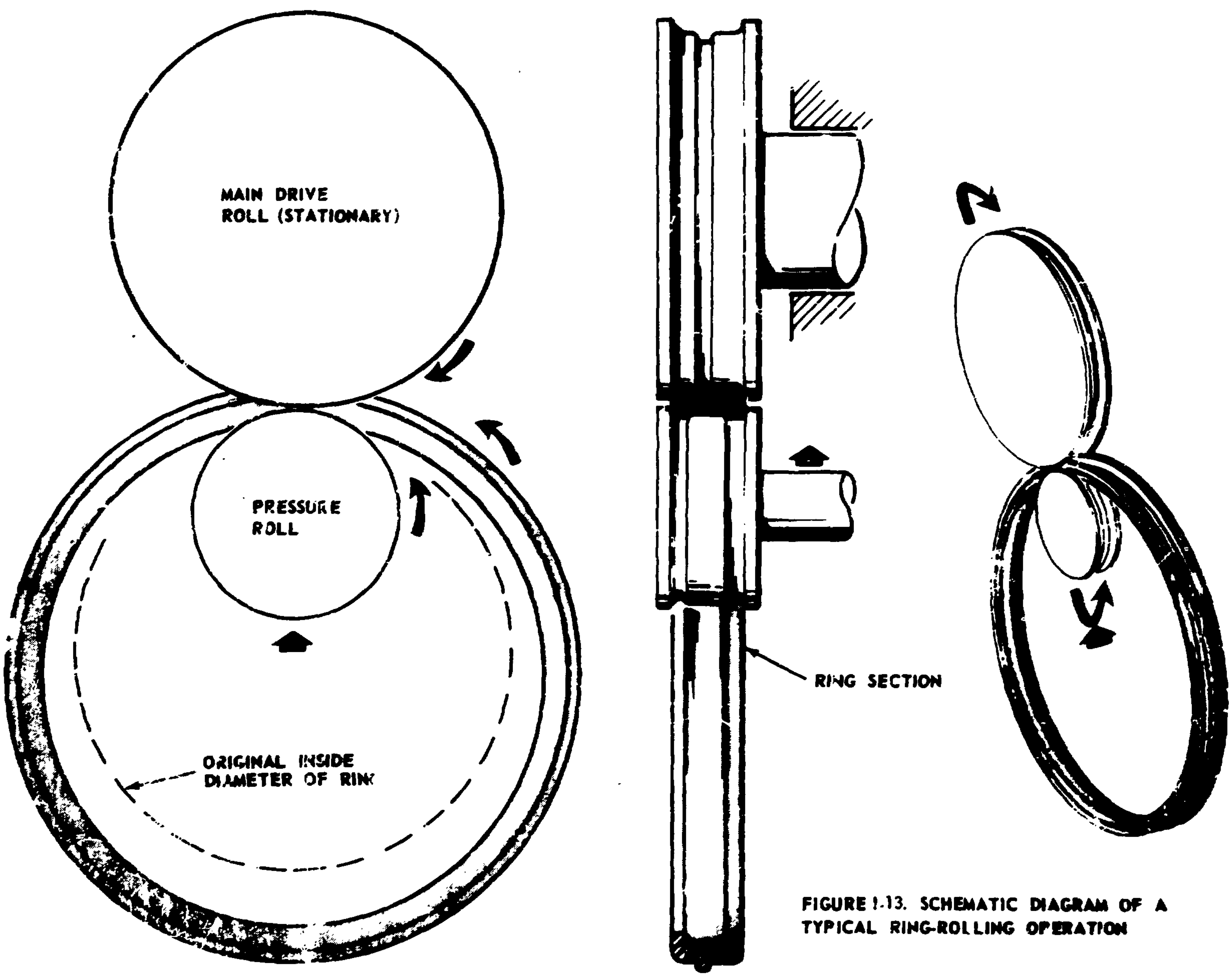


FIGURE 1-13. SCHEMATIC DIAGRAM OF A TYPICAL RING-ROLLING OPERATION

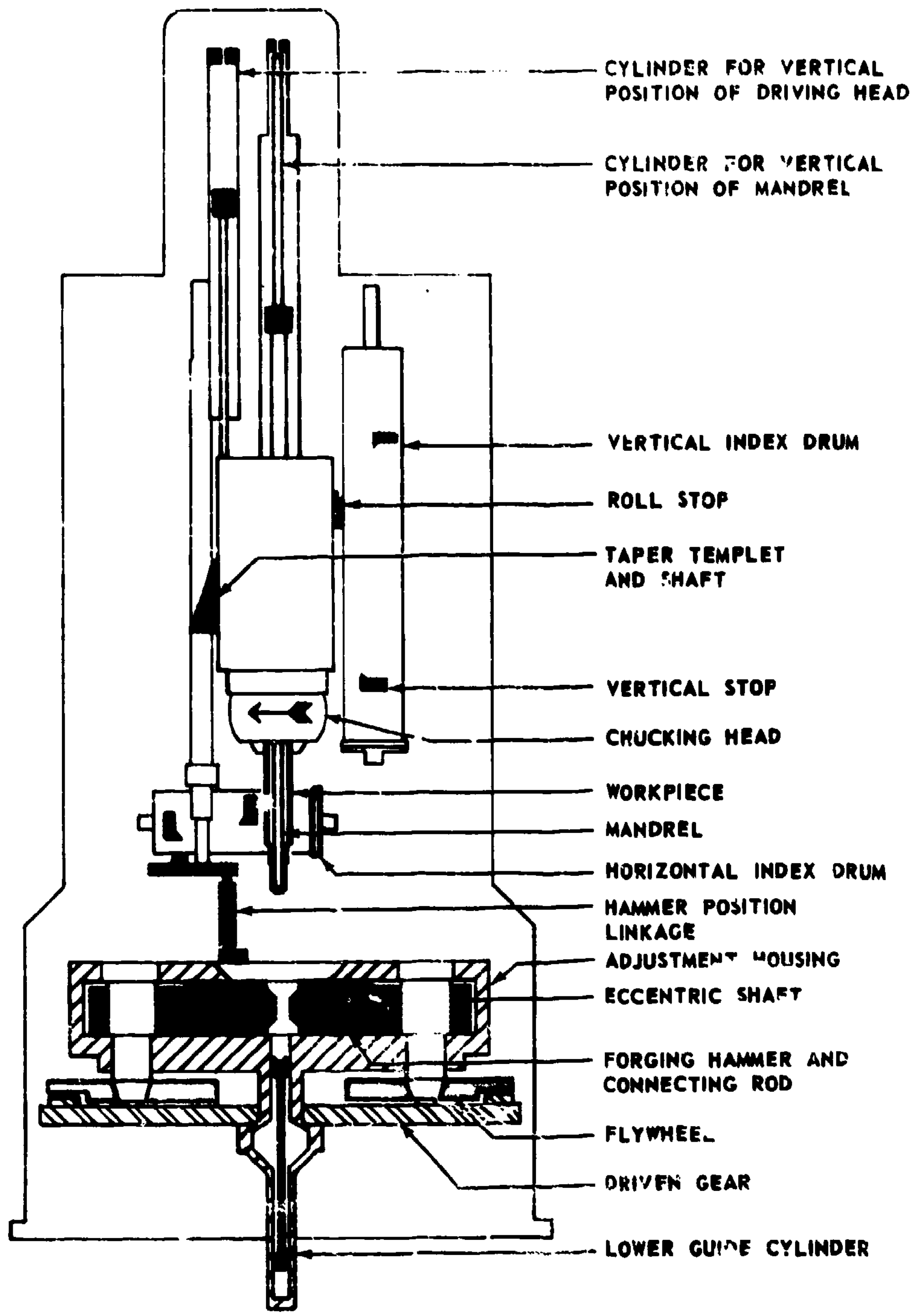


FIGURE 1.14. CUTAWAY DRAWING ILLUSTRATING THE BASIC OPERATING FEATURES OF A ROTARY FORGING MACHINE

held in a chucking head by jaws operated with compressed air. The chucking head rotates and guides the workpiece to the forging hammers. At each impact of the forging hammers, the workpiece is halted in its movement. An internal mandrel is used when tubing is being forged. Forging action is obtained through three or four small hammers that are connected to fast-moving eccentric shafts. Hammers are located at equal spacings around the work-piece, and all hammers impact at the same time. Thus shock transmission through the machine to the foundation is held to a minimum.

The machine is capable of completely automatic operation except for feeding and removal of the workpiece. The speed of forging may be changed manually during the forging cycle by the operator as desired. The dimensional accuracy and surface finish are a function of the chuck speed. Tolerances on tubing details can be held within ±0.012 inch on the OD and ±0.004 inch on the ID when forged on a mandrel. Tolerances for solid cylindrical forgings are ±0.008 inch for diameters below 2.35 inches and ±0.012 inch for larger diameters.

The rotary forging machine has a distinct advantage in costs because universal dies may be used for a large number of forging operations. The life of the dies, depending on operating conditions, is between 1000 to 2000 parts. Forging at elevated temperatures does not materially affect the life of the forging hammers since contact time of the workpiece is very short and sufficient cooling of the hammer is provided by the lubrication system and air blast.

Shear-Forming Machines

Shear forming, otherwise known as hydrospinning, spin forging, shear spinning, flow turning, etc., is a process that combines spinning with the ring-rolling process. It changes the shape of a plate or ring by applying force through pressure rolls operating against the workpiece, which is held by a turning mandrel. During the process, flat metal blanks are reduced in thickness between the rolls and the mandrel, and the shape is frequently changed to axisymmetrical shapes, such as cones, cylinders, and hemispheres. Although this operation is not considered forging, it is frequently tied in closely with other forging operations in the production of such shapes. Shear forming also is finding increasing application in the secondary shaping of rolled rings and forged hemispheres.

In one typical process, a blank is placed on the end of the mandrel and clamped by a hydraulic ram (Figure 1-15). A templet for the part is connected into a tracing system and the required pressure for the rollers is preset. The mandrel is then spun at a rate which normally exceeds 1000 surface feet per minute. A rate of speed is then established for the rollers. A feed-rate range from 0.005 to 0.150 inch per revolution is possible on most equipment. Slower feed rates give better surface finishes. Contact pressures on the rollers close to 400,000 psi are possible on current equipment.

During the process, the metal is always displaced parallel to the centerline of the part. This condition holds even though the forming takes place over a tapered cone because the material ahead of the forming tool is always parallel to the starting blank position. Since the material ahead of the rollers remains undisturbed, a lip of material with the original thickness may be obtained by stopping the process at any point. This condition also permits the use of a starting blank which has some shape other than circular.

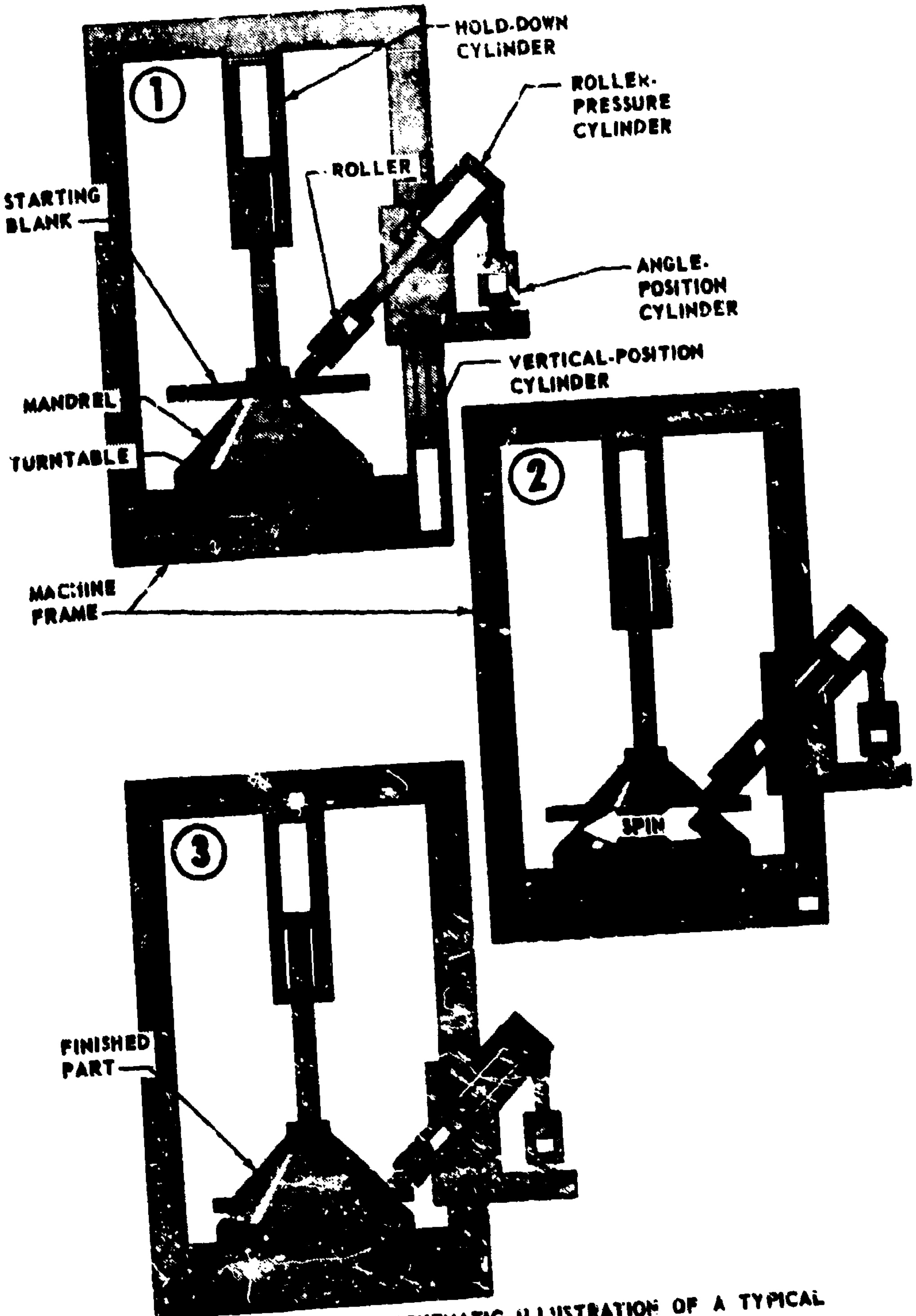


FIGURE 1.15. SCHEMATIC ILLUSTRATION OF A TYPICAL SHEAR-FORMING OPERATION

Most of the engineering materials of structural design can be shear formed, provided sufficient ductility is available.[4] Work has been performed on carbon, low-alloy and heat-resistant steels, superalloys, aluminum alloys, titanium alloys, magnesium alloys, and molybdenum alloys. However, when the materials lack sufficient ductility at room temperature they must be shear formed at elevated temperatures.

Tensile data are useful for predicting the maximum permissible reduction in spinning and shear forming.[5, 6, 7] For metals exhibiting a true necking strain of less than 0.5 in tension tests, the maximum spinning reduction improves with ductility. For metals with tensile reduction of area values above 45 per cent, the maximum spinning reduction in one pass is about 80 per cent and is relatively independent of the ductility value. The minimum included cone angle to which a metal may be shear spun from a flat blank is about 30 degrees. Present equipment capacity will produce a piece with a 70-inch maximum swing and a 50-inch maximum length.

The following shapes have been made on shear-forming equipment: (a) straight-wall cones, (b) straight-wall tubes, (c) curvilinear-wall shapes, and (d) hemispherical or elliptical shapes. For economy of operation, most of the curved shapes require a formed starting blank. The starting blank can be a sheet, plate, forging, casting, or formed and welded sheet.

The final wall thickness is determined by the starting-blank thickness and the included angle of the cone. The thickness at any point in the part measured parallel to the centerline of the part remains the same as the original thickness of the blank of the finished part. For this reason, thickness can be changed only by changing the original blank thickness or by changing the blank configuration. When shear forming a flat blank over a curved mandrel, the part thickness varies according to the angle tangent to the curvature at the point of contact. For this reason, a contoured blank is required to achieve a uniform wall thickness over a curved mandrel surface.

The following formula may be used to calculate the finished part thickness:

$$T_f = T \sin \alpha$$

where

T_f = the finished part thickness

α = the angle of the part measured from the centerline of the part

T = the starting-blank thickness.

Figure 1-16 illustrates a typical conical shape flow-turned from a blank of constant thickness, showing the location of the angle α and other values used in the above formula.

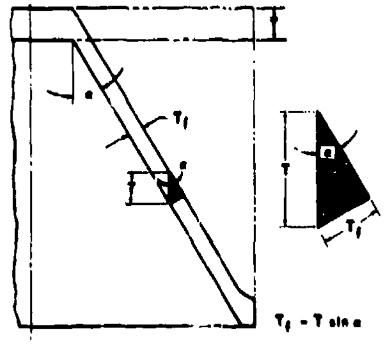


FIGURE 1.16. SKETCH ILLUSTRATING THE FINAL THICKNESS OF A CONE AS A FUNCTION OF THE MANDREL ANGLE

CHOICE OF HAMMERS OR PRESSES

Discussions concerning either drop-hammer forging or hydraulic-press forging generally cite certain advantages for one or the other methods. Essentially, press forging offers unique advantages over hammer forging only when faster deformation rates introduce new problems. For example, magnesium and beryllium are forgeable at press speeds, while at hammer speeds both metals often crack. Furthermore, castable alloys such as kirksite, beryllium-copper, and superalloys can be used for press-forging dies, but these materials usually lack the necessary shock resistance for hammer applications. The most significant disadvantages of press forging are confronted when the metal being forged requires considerably higher workpiece temperatures than can be achieved on the dies. Such metals are essentially "die quenched" during forging, that is, the temperature of the workpiece rapidly approaches that of the die. As a result, greater deformation pressures are necessary. This is particularly a problem if the metal becomes embrittled at some temperature between the original workpiece temperature and die temperature.

Metals forged in hammers are likely to exhibit a significant temperature rise during rapid deformation. This is a problem when forging metals like aluminum or superalloys at temperatures close to their melting temperatures. The temperature rise is usually less significant during press forging.

When forging steels and other alloys subject to scaling, the hammer has an advantage in that scale can be removed easily from the die cavity during forging and loosened from the workpiece by the repetitive striking action of the hammer. In presses, the scale is frequently pressed into the workpiece surface and is difficult to remove from die recesses.

Presses are favored for metals that have forging temperatures close to that achievable on dies. Conversely, hammers are preferred for forging metals that require higher workpiece temperatures. Presses are often preferred for converting cast ingots because it is easier to observe any cracks or ruptures that occur during forging.

TABLE 1-2. COMMENTS ON THE CHOICE OF EITHER HAMMERS OR PRESSES FOR FORGING ALLOYS OF SEVERAL METAL-BASE SYSTEMS

Alloy System	Choice of Equipment
Aluminum	Presses preferred if part requires severe deformation; otherwise optional
Beryllium	Presses preferred because of poor forgeability under rapid deformation rates
Copper	Hammers preferred because workpiece temperature is more easily maintained; presses preferred for forging bronze alloys and high-zinc brasses that are sensitive to deformation rate
Columbium	Hammer preferred for wrought alloys when high forging temperatures are required
Magnesium	Presses preferred because of poor forgeability under rapid deformation rates
Molybdenum	Hammers preferred for wrought alloys when high forging temperatures are required
Nickel	Choice depends on section-thickness requirements; hammers are preferred for forgings with thin sections (less than 1/2 inch); otherwise optional
Steels (carbon and low alloy)	Hammers preferred for forgings containing thin sections or when scale removal is a problem; otherwise optional
Steels (stainless grades)	Hammers preferred for forgings containing thin sections; otherwise optional
Tantalum	Hammers preferred for wrought alloys when high forging temperatures are required
Titanium	Hammers preferred for forgings containing thin sections; otherwise optional
Tungsten	Hammers preferred for wrought alloys when high forging temperatures are required.
Zirconium	Presses preferred if billets are jacketed (for temperatures above 1400 F); optional if forging temperature is below 1400 F

Table 1-2 lists several common forging-alloy systems and gives some general comments about the choice of either presses or hammers for forging the alloys. Because there is a basic difference in the time required to form a particular shape, the most practical forging design for hammers may differ from that for presses. The most noticeable difference is in section size for alloys having forging temperatures higher than that permissible for dies. There are a few cases where forging designs are particularly suited to one or the other forms of equipment. For example, multiple-ram presses have advantages in shape versatility not available in hammers. On the whole, however, the majority of materials and shapes can be forged interchangeably in either hammers or presses. In most cases, the choice is more a matter of economics than of technical considerations.

REFERENCES

(1) Naujoks, W., and Fabel, D. C., Forging Handbook, American Society for Metals, Cleveland, Ohio, First Printing (1939).

(2) Brochure from The Chambersburg Engineering Company, Chambersburg, Pennsylvania.

(3) Personal communication from Mr. M. Bohn of Curtiss-Wright Corporation.

(4) Ball, O., "Survey of the Shear Forming Process", North American Aviation, Report No. NA62H-29, (undated), Columbus, Ohio.

(5) Kegg, R. L., "A New Test for Determination of Spinnability of Metals", Journal of Engineering for Industry, Trans. ASME, Series B, 83, 119-124 (1961).

(6) Kalpakcioglu, Serope, "A Study of Shear-Spinnability of Metals", Journal of Engineering for Industry, Trans. ASME, Series B, 83, 478-484 (1961).

(7) Kalpakcioglu, Serope, "Maximum Reduction in Power Spinning of Tubes", ASME Preprint Paper No. 63-Prod 6 (1963).

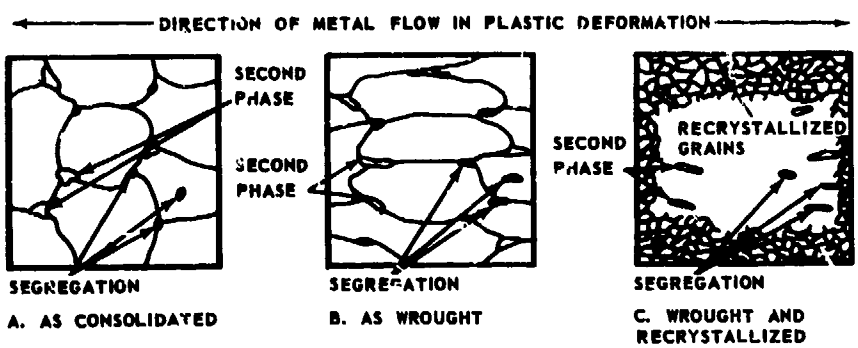
CHAPTER 2

METAL FLOW IN DIE FORGINGS

DEVELOPMENT OF GRAIN FLOW

As received for industrial forging operations, metals and alloys ordinarily exist in one of two basic forms: (1) consolidated (cast, or pressed and sintered) or (2) wrought (consolidated and deformed plastically).

Consolidated metals are made up of crystals or grains with random orientations, and exhibit essentially uniform properties in all directions. On the other hand, wrought material has been extruded, rolled, cogged, or otherwise elongated plastically, and usually exhibits better ductility in a direction parallel to that of the plastic elongation. This is particularly true of metals containing minor amounts of segregation or of a second phase. Through plastic deformation, the grains, the segregation, and any phases present become oriented parallel to the elongation. This orientation effect is called grain flow or fibering.



Grains, segregation, and second phase are randomly oriented.

Grains, segregation, and second phase are elongated in the direction of plastic deformation.

Grains randomly oriented. Segregation and second phase elongated in the direction of plastic deformation.

FIGURE 2-1. TYPICAL DEVELOPMENT OF GRAIN FLOW BY PLASTIC DEFORMATION OF METALS

The typical development of grain flow by plastic deformation is illustrated schematically in Figure 2-1, where the first sketch shows the random structure of a metal as it might appear after consolidating by either casting or by powder-metallurgy techniques. The second sketch shows how the metal structure is oriented in a particular direction by plastic deformation. The third sketch shows that, even after recrystallization, the segregation and second phase are oriented along the axis of principal deformation.

Not all of the effects of grain flow are this obvious. For example, cast alloys are usually heterogeneous and exhibit localized alloy segregation. After deformation, the alloy-rich and alloy-lean areas form alternating layers parallel to the direction of principal elongation. This effect, which is illustrated in Figure 2-2, is called banding.

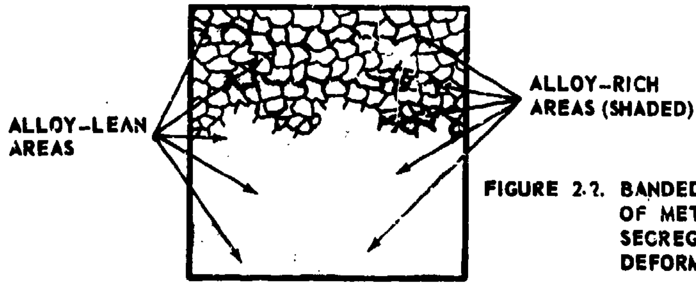


FIGURE 2-2. BANDED STRUCTURE TYPICAL OF METALS CONTAINING ALLOY SEGREGATION AFTER PLASTIC DEFORMATION

There are many factors that can contribute to grain flow effects. Several of these factors and their general influence on properties are:

Ductile Segregation.

Segregation that deforms plastically with the base metal will become elongated during forging. This change in shape causes ductility and toughness to improve in the longitudinal direction* and to worsen in the transverse direction* compared with these properties of the starting billet. (Example: sulfides in steel.)

Brittle Segregation.

Segregation that does not deform plastically with the base metal will usually break up to form discontinuous bands of smaller particles. (Examples: undissolved carbides in stainless steels; nitrides in superalloys.) The ductility of metals containing brittle segregation will usually improve in both the longitudinal and transverse directions with increasing plastic deformation.

Alloy Segregation.

Alloys containing inhomogeneous-alloy distributions will develop a banded structure after plastic deformation. Banding in this case consists of alternating layers of alloy-rich and alloy-lean areas The principal mechanical properties affected by banding are strength and ductility. Higher values for both properties are generally found in the longitudinal direction. (An example of alloy segregation in as-cast material is found in dendritic structures, where the cores of the dendrites are of a slightly different alloy composition than the surrounding material). Alloys given increasing amounts of hot work will usually become more homogeneous, and eventually the effect of alloy segregation will become negligible.

*The longitudinal direction is parallel to the major direction of plastic elongation. A wrought bar, for example, has longitudinal grain flow parallel to the bar axis. Conversely, the transverse direction is perpendicular to the direction of plastic elongation.

Low-Melting Phases.

When metals are consolidated by casting, the alloy segregation that occurs during solidification usually results in the formation of low-melting phases (e. g. eutectics, sulfides, etc.). Such phases solidify in the original grain boundaries and may cause weaknesses at forging temperatures. In many instances, these phases can be dissolved in the base metal by suitable homogenization treatments at temperatures just below their melting temperatures. This usually improves forgeability. If the low-melting phases cannot be eliminated by this treatment, they generally influence directional properties as does either ductile or brittle segregation. If dissolved, the low-melting phases influence directional properties like alloy segregation does.

Second Phases.

For reasons determined by their composition and constitutional diagrams, certain alloys ordinarily contain two phases. (Examples: ferrite and austenite in high-chromium stainless steels, alpha and beta phases in some titanium alloys.) When the two phases are elongated by plastic deformation, they cause directional variations in properties. The shape of the elongated grains, however, can usually be altered by heat treatments, thus minimizing directional variations. By annealing the alpha-beta titanium alloys at certain temperatures, the alpha phase becomes equiaxed. This change in grain shape eliminates most of the directionality in mechanical properties.

Crystalline Anisotropy.

When metals are deformed, their grains often become oriented in such a fashion that there is a distinct alignment of certain crystallographic planes. This removes the randomness ordinarily associated with polycrystalline materials and promotes a condition favorable for anisotropic mechanical properties.

Die forgings are seldom forged directly from as-consolidated materials. Usually, they are forged from wrought materials that have been either rolled, extruded, or cogged to bar shape. Sometimes, as-cast ingots or pressed-and-sintered bars are upset forged between either flat or generously contoured dies, but these are steps preliminary to die forging with the purpose of developing favorable wrought structures in shapes that ultimately are to be forged to disk-type configurations.

Figures 2-3 and 2-4 show schematically how grain flow in the original bar stock is reoriented during the forging of two types of die forgings: a disk and a rib-and-web shape. A series of loops is developed in the grain-flow pattern of the disk forging. These grain-flow loops are developed during the initial upset forging operation and carry through the final operation. By contrast, the grain flow in the rib-and-web forging forms a continuous pattern that conforms essentially to the part outline.

The grain-flow patterns shown in Figures 2-3 and 2-4 indicate that metal flows in various directions; the extent is determined by the forging die and the type of forging

operation. The principal flow of metal in disk forgings is in the radial direction. However, by virtue of the increasing diameter there is considerable elongation, or flow, in the tangential direction. The terms applied to these types of metal flow are radial grain flow and tangential grain flow, respectively. The metal flowing upward into the rim is called vertical grain flow. Terms of similar derivation apply to the rib-and-web forging, where longitudinal grain flow refers to flow in the direction of the original billet axis, long-transverse grain flow refers to flow at right angles to the longitudinal grain flow and parallel to the plane of the die, and short-transverse grain flow refers to flow at right angles to both the plane of the die and to the longitudinal grain flow.

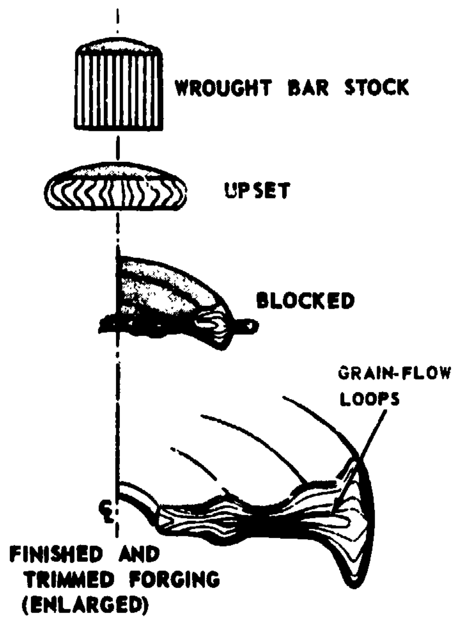


FIGURE 2-3. TYPICAL SEQUENCE OF FORGING STEPS FOR A DISK SHAPE, SHOWING THE DEVELOPMENT OF GRAIN FLOW

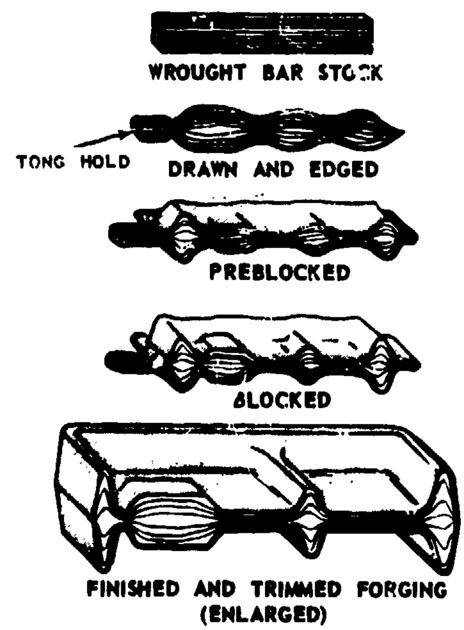
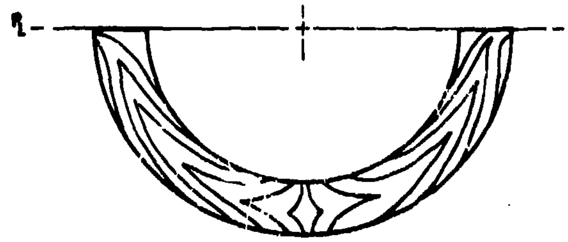


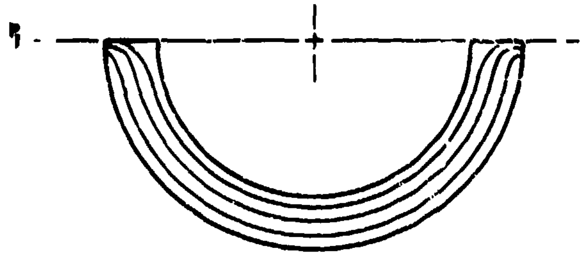
FIGURE 2-4. TYPICAL SEQUENCE OF FORGING STEPS FOR A RIB-AND-WEB SHAPE, SHOWING THE DEVELOPMENT OF GRAIN FLOW

Control of Grain Flow

For certain applications it is desirable to control the grain flow developed during die forging. The forging of steel hemispherical parts, for example, often includes a preliminary side-forging step instead of the conventional upsetting common to most other circular parts. Figure 2-5 compares two grain-flow patterns developed in hemispherical parts. The lower sketch shows how smooth, unbroken flow lines result from a preliminary side-forging step. The upper sketch shows how the weaker center material is concentrated in the polar region by normal forging practices.



A. FORGED HEMISPHERE, UPSET BEFORE DIE FORGING, SHOWING WEAKNESS IN POLAR REGION



B. FORGED HEMISPHERE, SIDE FORGED BEFORE DIE FORGING, SHOWING NO WEAKNESSES IN POLAR REGION

FIGURE 2-5. COMPARISON OF FLOW LINES DEVELOPED IN HEMISPHERICAL FORGINGS WITH DIFFERENT PRELIMINARY FORGING TECHNIQUES

Most methods for altering grain flow are not this simple. As stated earlier, inclusions ordinarily line up with grain flow; if the grain-flow pattern is straight, so are the inclusions. Because cylindrical parts forged from wrought bar will retain this basic pattern, it is sometimes desirable to alter the forging practice to break up the straight-line pattern in such parts. This reduces the probability of long inclusions appearing on the surface of finished parts and causing rejections. Forge shops will sometimes use a corrugated die design to provide a wavy grain-flow pattern for this purpose. This practice, illustrated by Figure 2-6, has been known to reduce rejections from nearly 30 per cent to less than 5 per cent on parts of cylindrical shape.

Forging and lubrication techniques can also be used for controlling grain flow. If, for example, a disk is hammer forged between contoured dies, the top surface of the forging is generally lubricated better than the lower surface. This can cause the type of nonuniform grain flow illustrated in Figure 2-7. Essentially, the metal flows readily against the upper-die surface but is virtually undeformed near the lower-die surface. In addition to inadequate lubrication, die-chilling effects contribute to this type of flow. The out-bent fibers formed in this way result in poorer toughness and surface finish. The condition can be corrected by turning the disk over during forging or by providing better lubrication on the lower die surface. A third alternative consists of upsetting between flat dies to a disk slightly smaller than the final die, turning the disk over, then finishing in the impression die. This latter method is the one most often used in normal forging practice because the undersized flat disk is easier to locate in the die and the grain-flow patterns are generally more uniform. The flow patterns can be adjusted in other shapes by similar lubricating and forging techniques

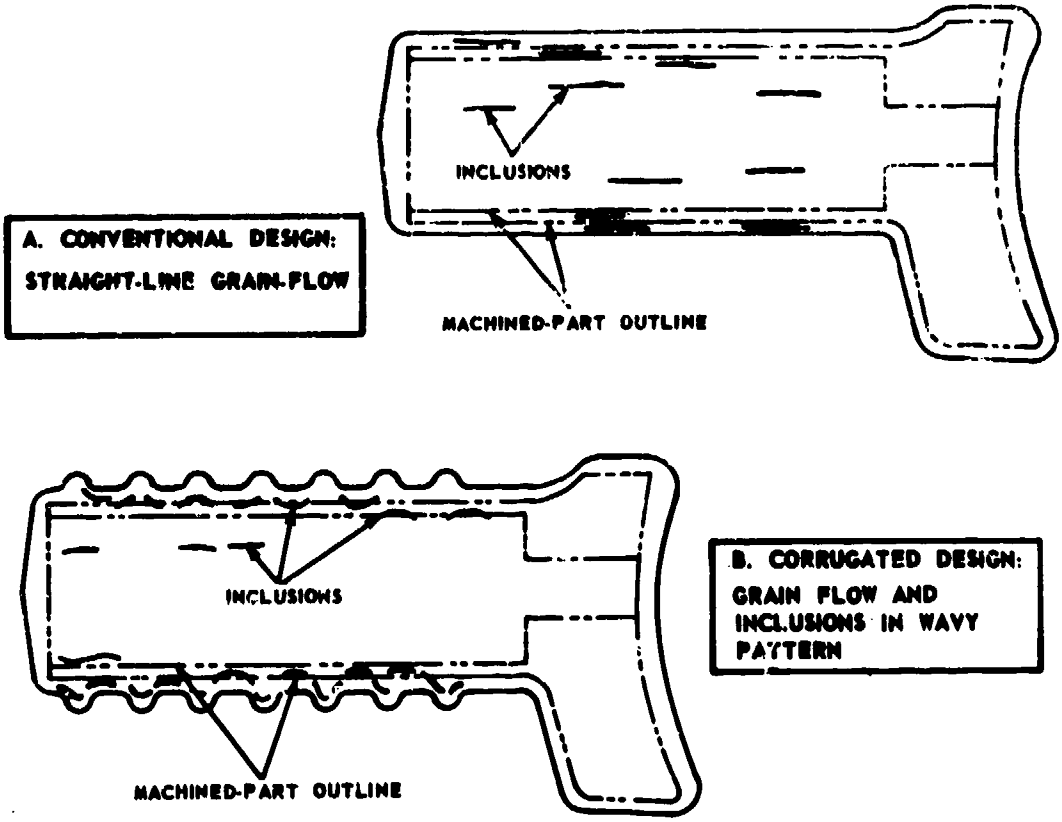


FIGURE 2-6. COMPARISON OF DESIGNS SHOWING HOW STRAIGHT-LINE GRAIN-FLOW PATTERNS IN CYLINDRICAL FORGINGS CAN BE BROKEN UP BY USING A "CORRUGATED" DIE DESIGN

FIGURE 2-7. NONUNIFORM FLOW PATTERN IN A DISK FORGED WITH GOOD LUBRICATION ON TOP SURFACE AND POOR LUBRICATION ON BOTTOM SURFACE

METAL FLOW IN IMPRESSION DIES

As described earlier, regions within a forging that exhibit the most pronounced flow patterns are generally those that have received the greatest deformation during forging. It is a significant fact that the amount of deformation is seldom uniform throughout a forging. It is not uncommon, for instance, to find both severely deformed regions and "dead-metal" zones within a forged shape. A vivid illustration of this is shown in Figure 2-8.

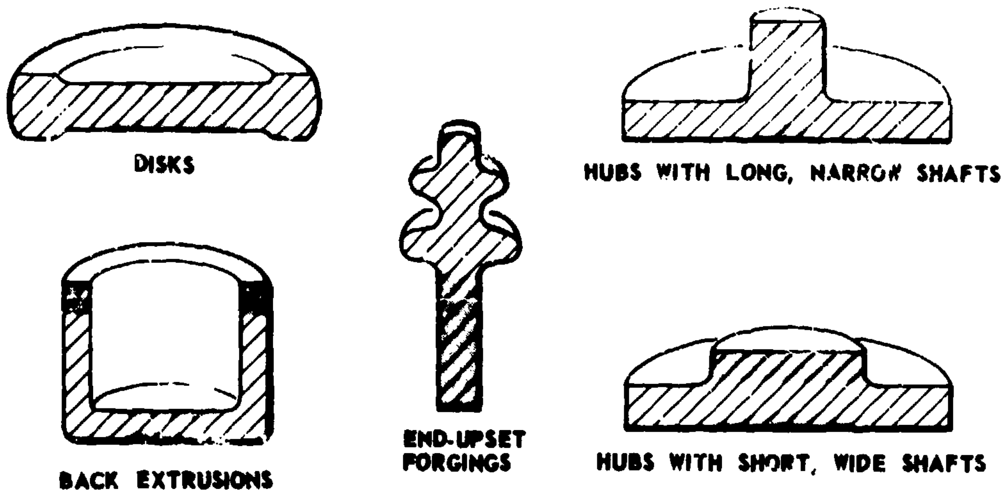
FIGURE 2-8. GRAIN-FLOW PATTERN OF A DISK FORGING, SHOWING BOTH "DEAD-METAL" ZONES AND SEVERELY DEFORMED REGIONS

In die forgings, the regions of minimal flow, or dead-metal zones, are often called "die-locks". Figure 2-9 shows where the dead-metal zones normally appear within die forgings of various shapes. These examples are further illustrations of how nonuniform the deformation is during most die-forging operations.

In the case of alloys that are to be hot-cold worked, it is often desirable to design forgings in such a way that the dead-metal zones are removed during machining. If this is not practical, preliminary forging steps may be used to improve the uniformity of deformation during the final operation. Methods for designing such preliminary dies for several typical shapes are illustrated in Figure 2-10. When these preliminary shapes are forged subsequently in the finishing die, deformation is provided in regions that otherwise would be dead-metal zones in conventionally forged parts. Even with such techniques, however, the amount of deformation is seldom homogeneous.

Effects of Nonuniform Deformation

The need for uniform, controlled deformation varies with the material being forged. In the case of alloy steels, uniform deformation is not essential. In the case of superalloys, the total amount of deformation is not very important, but the amount imparted during the last forging step significantly influences the grain size developed by subsequent heat treatment. A good illustration of variations in grain size in a heat-treated superalloy forging is shown in Figure 2-11. This part was forged in a worn set of dies and then restruck for sizing in a new set. During the final sizing operation, the forging was deformed only a small amount, a practice that makes metals susceptible to abnormal grain growth. It is usually desirable to produce enough deformation in the final forging operation to insure recrystallization to a fine grain size.



SHADED AREAS DENOTE NORMAL LOCATIONS OF DEAD-METAL ZONES

FIGURE 2-9. USUAL LOCATIONS OF DEAD-METAL ZONES IN SEVERAL TYPICAL SHAPES FORGED BY CONVENTIONAL PRACTICES

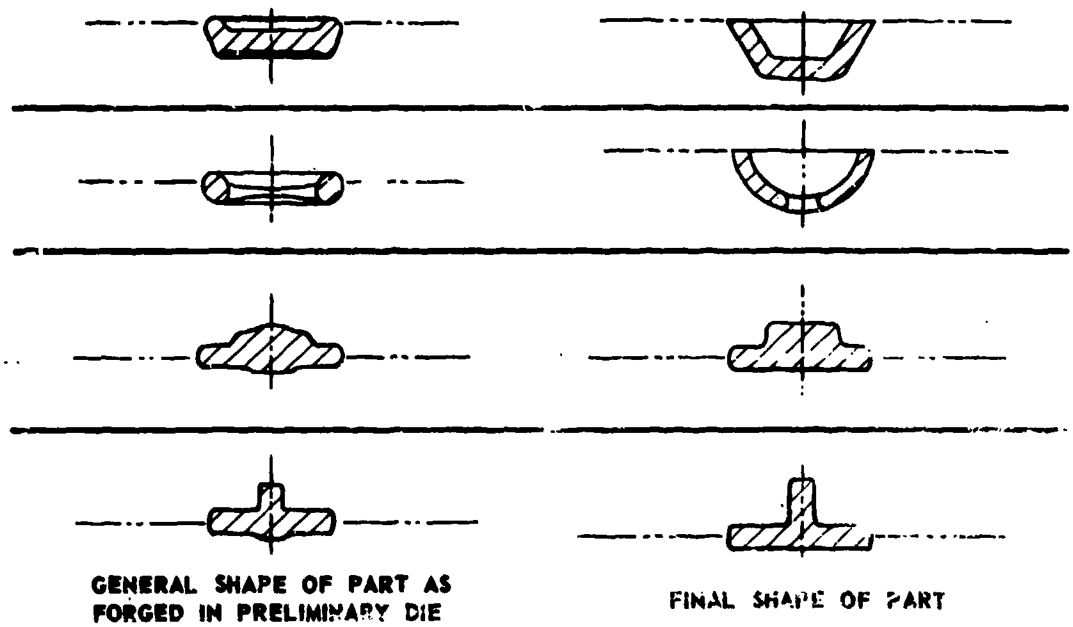


FIGURE 2-10. METHODS FOR DESIGNING PRELIMINARY DIES TO OBTAIN MORE UNIFORM DEFORMATION IN TYPICAL FORGED SHAPES

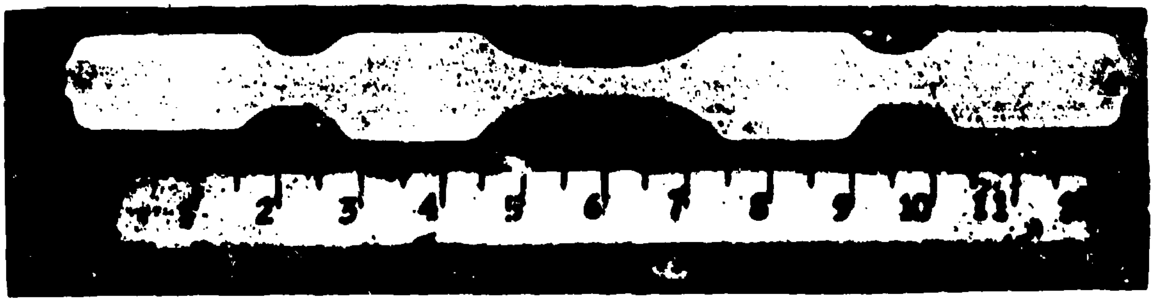


FIGURE 2-11. ETCHED CROSS SECTION OF A HEAT-TREATED SUPERALLOY DISK FORGING, SHOWING GRAIN-SIZE VARIATION CAUSED BY INSUFFICIENT REDUCTION IN THE LAST FORGING OPERATION

Alloy: D979

Finish forged at 1950 F in worn die set and hot sized in new die set.

Solution heat treated at 1900 F, 2 hours; oil quenched.

Certain metals forged by hot-cold working schedules, such as molybdenum alloys and some stainless steels (e. g., 19-9DL, 16-25-6), require reductions controlled between both maximum and minimum levels. In such materials, the appropriate forging reduction varies with the strength desired. In the case of the titanium alloy B-120VCA, careful control of deformation schedules improves the response to subsequent aging treatments.

For these reasons, forging schedules for some alloys should be designed to produce appropriate amounts of deformation in the final operation. In many instances, this means that preliminary or intermediate dies should be used even though they would not be required for meeting purely geometrical requirements.

To illustrate this point, Figure 2-12 shows the forging operations and die sequences ordinarily needed to make a typical disk shape from each of two different materials: a low-alloy steel (4340) and a hot-cold work stainless steel (16-25-6). The shapes and forging temperatures shown are considered typical for the two alloys. Although the 16-25-6 alloy can be forged in Step 2 into a shape similar to that illustrated for 4340, the generously contoured shape is preferred for controlling the final amount of deformation and, thus, the metallurgical and mechanical properties. As indicated, the stainless steel is ordinarily upset and blocked at a hot-working temperature. Then the blocked shape is given a controlled deformation in the final die at a cold-working temperature (1500 F or lower). For this type of forging, the blocked shape is usually about 20 per cent smaller in diameter than the final die to insure uniform deformation. Such shapes are sometimes machined to permit even closer control of reductions. The finishing die is usually designed with a comparatively thick web so that the dead-metal surface layer can be removed when the part is finished by machining.

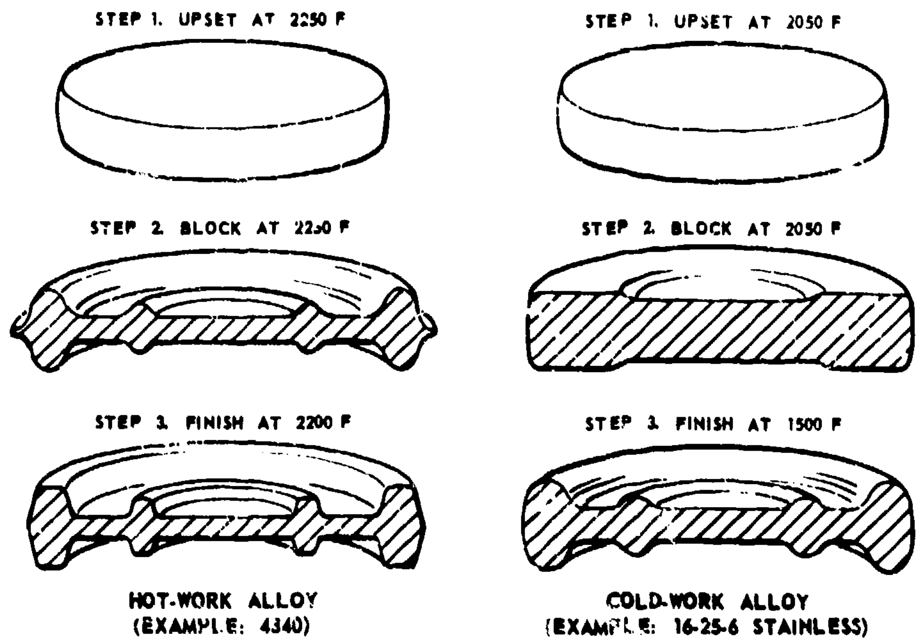


FIGURE 2-12. FORGING AND DIE SEQUENCES TYPICAL FOR HOT-WORK ALLOYS AND HOT-COLD WORK ALLOYS

The same principles apply to the forging of refractory metals except that the preliminary deformation is ordinarily conducted at cold-working temperatures. Table 2-1 indicates the relative importance of controlled deformation during die forging of different types of metals. These points are discussed more thoroughly in later chapters covering each alloy system.

TABLE 2-1. THE IMPORTANCE OF CONTROLLED FORGING DEFORMATION WHEN FORGING SEVERAL ALLOYS

Alloy Group and Examples	Type of Forging Cycle	Need for Controlled Deformation
Carbon and alloy steels (1035, 4340)	Hot work	Level of deformation [illegible] important for final properties
Superalloys (A286, Inco 901)	Hot work	Level of deformation must be large enough to avoid abnormal grain growth; above certain minimum levels, the amount is not essential
Austenitic stainless steels (16-25-6, 19-9DL)	Hot-cold work	Level of deformation in final die requires control and uniformity
Refractory metals (unalloyed molybdenum)	Cold work	Level of deformation accumulated during all forging steps requires control and uniformity

Estimating the Amount of Deformation

It is frequently desirable to estimate the amount of plastic deformation resulting from a particular forging practice or to estimate the strains at different locations within the workpiece. Such estimates are useful in identifying the severity of deformation producing satisfactory properties and in converting laboratory results to full-scale operations. Many authors use the terms "forging ratio", "upsetting ratio", or "reduction in area" to describe the severity of plastic deformation. They are ordinarily defined as

$$\text{Forging Ratio} = \frac{\text{Original Area}}{\text{Final Area}}$$

(Both areas are measured in a plane perpendicular to the principal direction of flow or fiber direction)

$$\text{Upset Ratio} = \frac{\text{Original Height}}{\text{Final Height}}$$

$$\text{Reduction in Area, \%} = 100\left(\frac{\text{Original Area - Final Area}}{\text{Original Area}}\right)$$

(Both areas are measured in a plane perpendicular to the principal direction of flow)

Unfortunately, such terms are ambiguous and can be misleading unless geometric similarity is maintained between the original and final shapes. The ambiguity results from the fact that metal flow usually occurs in several directions, as indicated in Figure 2-13.

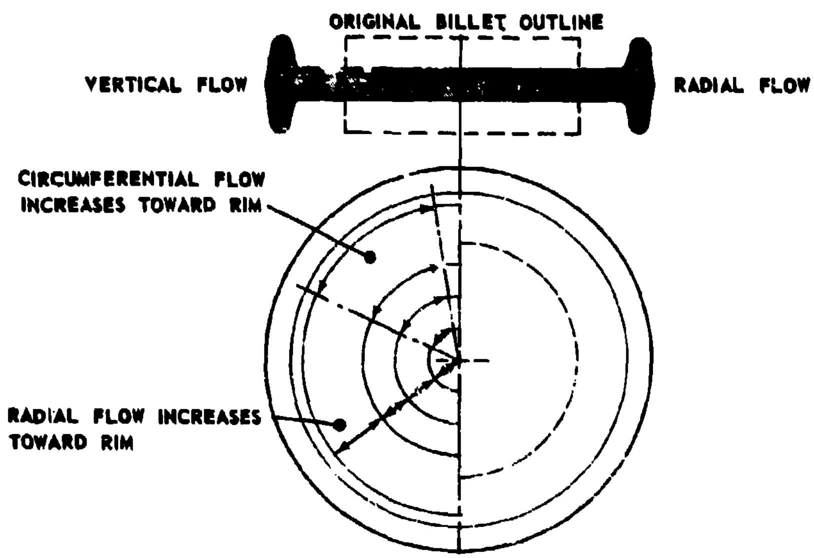


FIGURE 2-13. AN ILLUSTRATION OF THREE DIRECTIONS OF METAL FLOW DURING FORGING OF DISK SHAPE

It is usually more useful to describe the plastic strain in a body by the three principal deformations (e_1, e_2, e_3) in mutually perpendicular directions. For a spherical body deformed to an ellipsoid, the three principal deformations are in the directions of the three axes of the ellipsoid. The principal deformations are calculated as follows:

$$e = \frac{\text{Final length - Original length}}{\text{Original length}} = \frac{l_f - l_o}{l_o}$$

$$e = \frac{\text{Final height - Original height}}{\text{Original height}} = \frac{h_f - h_o}{h_o}$$

$$e = \frac{\text{Final width - Original width}}{\text{Original width}} = \frac{w_f - w_o}{w_o}$$

The subscripts (1, 2, 3) are assigned according to sign and magnitude of the deformations. Arbitrarily, e_1 is the largest strain and is always positive in sign; e_3 is the smallest strain and is always negative, e_2 can be either positive or negative in sign. Thus, $e_1 > e_2 > e_3$. For upsetting operations, the principal deformation, e_1, is numerically equal to the forging ratio and to the reciprocal of the upset ratio. For operations producing extension principally in one direction (extrusion, rolling and drawing out by forging), e_1 is positive.

Since plastic deformation does not change the density of a metal significantly, the original volume (V_o) and final volume (V_f) of the workpiece are identical. This leads to the following relationships:

$$V_o = V_f = V_o\,(1 + e_1)\,(1 + e_2)\,(1 + e_3)$$

and

$$1 = (1 + e_1)\,(1 + e_2)\,(1 + e_3) \quad ,$$

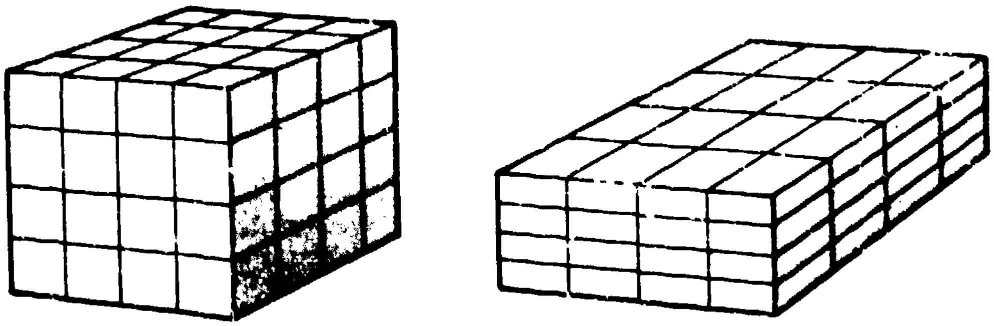
or

$$0 = \ln\,(1 + e_1) + \ln\,(1 + e_2) + \ln\,(1 + e_3) \quad .$$

The latter equation indicates that at least one of the "effective deformations", $\epsilon = \ln (1 + e)$, must be positive, another negative, and the third can be either. The single effective deformation value equal to the sum of the other two, but differing in sign, is called the maximum strain (ϵ_{max}). It is numerically equal to one-half of the sum of the individual deformations, neglecting the algebraic signs.

Figure 2-14 illustrates this system of describing and calculating the average amounts of deformation in an upsetting operation. In this example, the principal strains differ with direction because geometrical similarity was not maintained on the cross section. The maximum strain (ϵ_{max}) occurs in the height direction and is negative in sign. The largest positive value for effective deformation is in the length direction because the proportional extension was greater in length than in width.

Similar calculations can be made on other shapes and shape changes by selecting appropriate dimensions for changes in length, width, and height. The approach is especially useful for estimating differences in severity of deformation in various parts of a forging. For such purposes, the dimensional changes are often measured from



INITIAL FINAL

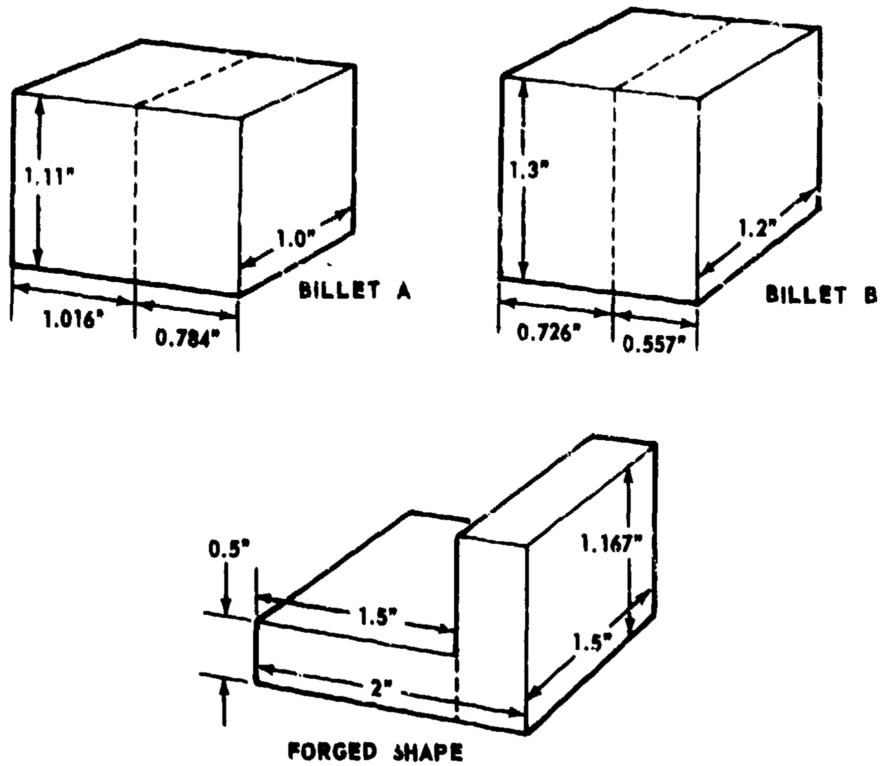
Example

	Height	Width	Length
Initial	1.0"	1.0"	1.0"
Final	0.5"	1.2"	1.667"

Direction of Strain	Principal Deformation, e	1 + e	Effective Deformation (ϵ), ln (1 + e)
Height	$\frac{0.5-1}{1} = -0.5 = e_3$	0.5	−0.6932
Width	$\frac{1.2-1}{1} = +0.2 = e_2$	1.2	+0.1823
Length	$\frac{1.667-1}{1} = +0.667 = e_1$	1.667	+0.5109

ϵ_{max} = Maximum strain = −0.6932

FIGURE 2-14. SCHEMATIC ILLUSTRATION OF UNIFORMLY UPSET PRISM AND CALCULATION OF PRINCIPAL DEFORMATIONS AND STRAINS



DASHED OUTLINE SHOWS CORRESPONDING VOLUMES OF BILLETS AND FORGED SHAPE.

	Web	Rib
As Forged from Billet A:		
Nominal value of ϵ_{max}	−0.799	−0.451
Direction	height	width
As Forged from Billet B:		
Nominal value of ϵ_{max}	−0.957	+0.223
Direction	height	length

FIGURE 2-15. EFFECT OF BILLET SHAPE ON MAXIMUM STRAINS DEVELOPED IN RIB-AND-WEB SECTIONS OF A FORGING

grids marked on appropriate interior surfaces of samples assembled from several pieces for the deformation studies.

Sometimes even rough estimates are helpful. Figure 2-15 shows that even a small change in billet dimensions can influence the amount and uniformity of deformation. Forging a rib-and-web section from either of the two indicated billets results in more severe deformation in the web section. However, forging from Billet A gives more deformation in the rib and less in the web than forging from Billet B. Thus, a forging produced from Billet A would be expected to have a more uniform response to heat treatment and less directionally in ductility values.

The schematic sketch in Figure 2-15 does not take into account the nonhomogeneous deformations characteristic of real forgings. Figure 2-16 illustrates a few important details. Ribs are ordinarily tapered in order to provide draft, and deformations increase in severity toward the narrower extremity. Metal close to the punch is ordinarily strained more than metal closer to the exterior. As indicated in Figure 2-16, metal flows laterally to form flash as well as vertically to fill the ribs. The two most severe regions of flow are those in the plane of the flash and adjacent to the inside fillets.

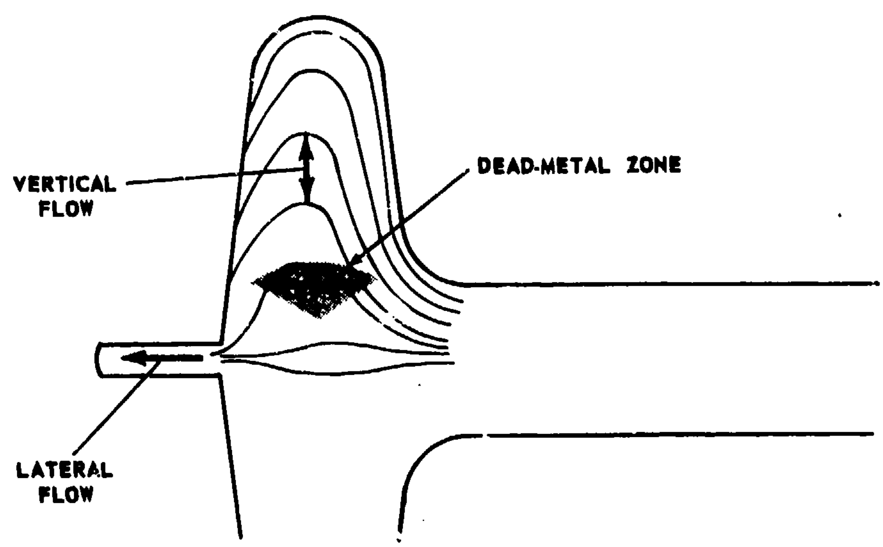


FIGURE 2-16. TYPICAL GRAIN-FLOW PATTERN IN A VERTICAL RIB

The shaded area is deformed the least and can be considered a dead-metal zone. Usually the amount of deformation in such regions can be increased, and uniformity improved, by controlling the shape of the preform or preliminary forging. The technique illustrated in Figure 2-17 can be used to increase forging reductions in the dead-metal zone to about 25 per cent in area. Unless a bead is used at the parting line, defects are apt to occur in the web from upsetting.

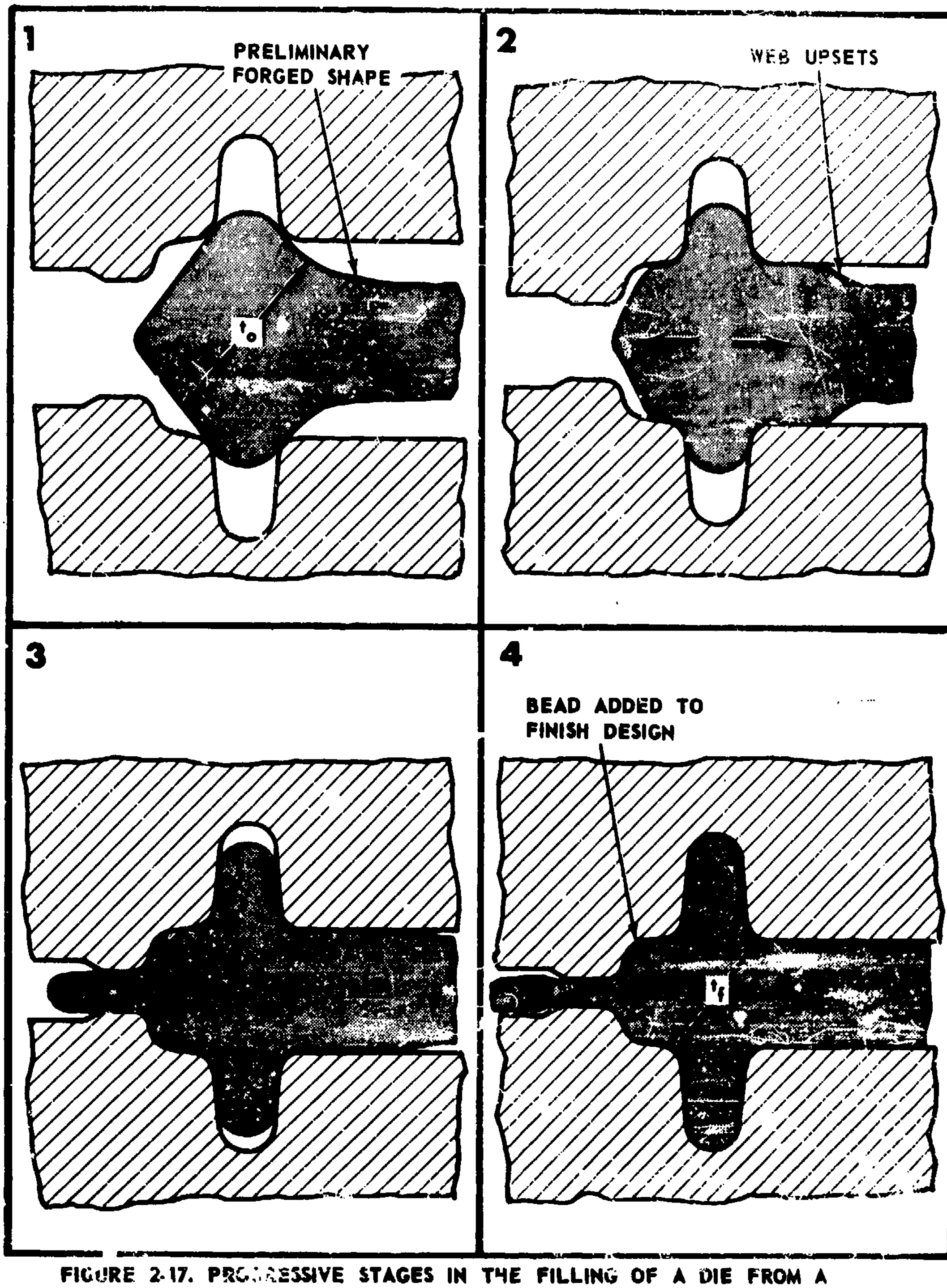


FIGURE 2-17. PROGRESSIVE STAGES IN THE FILLING OF A DIE FROM A SPECIAL PRELIMINARY SHAPE DESIGNED TO PROVIDE GREATER REDUCTION IN THE OTHERWISE DEAD-METAL ZONE OF A FINISHED RIB-AND-WEB FORGING

FACTORS INFLUENCING DIE FILLING

Many variables influence the ability of metals to flow during forging and to fill the cavities of impression dies. Some of the more important factors can be classified as follows:

(1) Forgeability and flow strength

(2) Friction and lubrication

(3) Die temperature

(4) Shape and size factors.

Forgeability and Flow Strength

Difficulties in filling die cavities are encountered in forging metals characterized by either poor forgeability or high flow strengths. Poor forgeability causes rupture before the die is filled; high flow strengths may cause metals to flow past recesses without filling them or may result in underfilling with the maximum loads available. Figure 2-18 suggests, in terms of forgeability and flow strength, the ease with which different metals and alloys can be deformed to fill intricate die cavities. Some of the difficulties attributed to these characteristic differences among materials can be illustrated by considering two particular types of forgings.

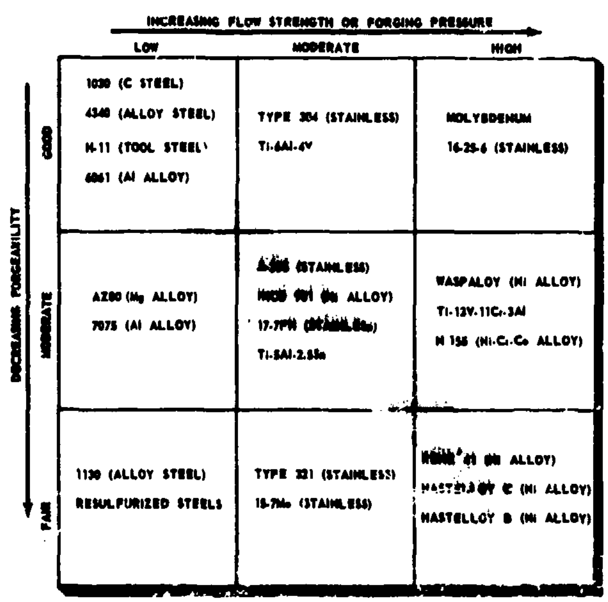


FIGURE 2-18. GENERALIZED DIAGRAM OF INFLUENCE OF FORGEABILITY AND FLOW STRENGTH ON DIE FILLING

Connecting Rods. Table 2-2 provides data obtained in experimental hammer forging of several alloys in a set of dies ordinarily used for 1030 steel connecting rods. As would be expected, the stronger materials required more reheating operation and a larger number of blows. Even so, the drawing and fullering* operations, ordinarily employed to achieve the proper stock distribution for impression forging, were unsuccessful with the two nickel-base superalloys. Figure 2-19 shows four of the specimens after forging. The forgeability of the higher alloy stainless and the iron-base superalloy was good enough to avoid cracking, but the hammer was too small to fill the dies completely.

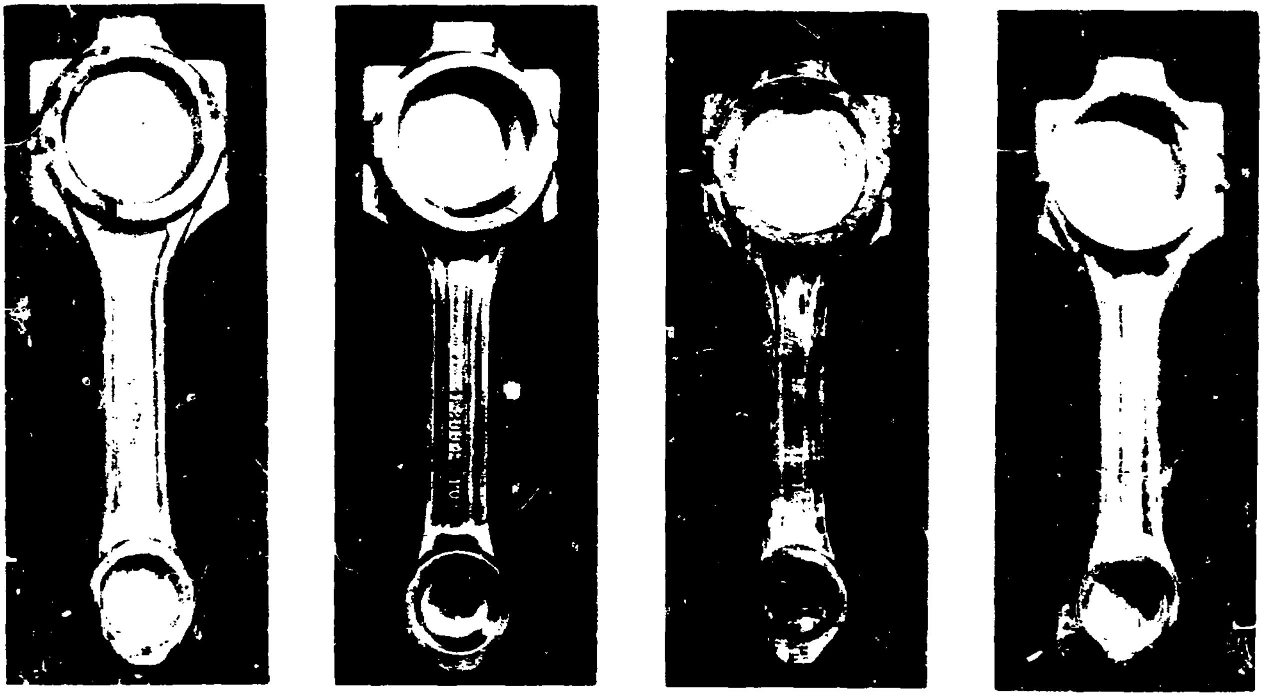
TABLE 2-2. FORGING DATA ON CONNECTING RODS FORGED FROM EIGHT DIFFERENT ALLOYS REPRESENTING INCREASING FORGING DIFFICULTY

Forging conducted on a 3000-pound hammer

Alloy	Forging Temperature, F	Number of Times Heated	Total Number of Blows to Complete Forging	Remarks
1030	2300	1	≤20	Completely filled
Type 304	2200	1	40	Completely filled
Type 347	2200	1	50	Completely filled
Monel	2050	2	51	Completely filled
16-25-6	2000	5	76	Incomplete fill on shaft end
N-155	2[illegible]0	5	89	Incomplete fill on both ends and ribs
Hastelloy B	2200	3	--	Extreme difficulty in drawing and fullering shape; experiment discontinued
Hastelloy C	2150	4	--	Same as for Hastelloy B

Hub Forgings. Table 2-3 gives data from a similar study on seven metals forged in dies ordinarily employed for forging hubs from 4340 steel. Both ruptures and pronounced underfilling occurred in the stronger nickel-base superalloys with poor forgeability. Sound hubs were made from the other four alloys but the dies were not completely filled. The degree of underfilling appears to depend on the flow strengths of the alloys at the forging temperatures used.

*In fullering, the cross-sectional area of a bar is reduced along the mid-section between the ends of the bar.



A. Type 304 Stainless

Completely filled.

B. Monel

Completely filled.

C. 16-25-6 Stainless

Incomplete fill on large end.

D. N-155 Superalloy

Incomplete fill on ends and ribs.

Increasing Forging Pressure →

FIGURE 2-19. APPEARANCE OF PARTS FORGED IN CONNECTING-ROD DIES FROM FOUR ALLOYS REPRESENTING INCREASING FORGING DIFFICULTY

Poor die fill indicated by arrows

TABLE 2-3. FORGING DATA ON HUBS FORGED FROM SEVEN DIFFERENT ALLOYS REPRESENTING INCREASING FORGING DIFFICULTY

Alloy	Forging Temperature, F	Number of Times Heated	Number of Hammer Blows Required to Complete Forging				Remarks[a]
			Upset	Block	Finish	Total	
4340	2300	1	2	4	5	11	Sound; completely filled
Monel	2100	1	2	5	7	14	Sound; shaft underfilled by 9/16 inch
Type 304	2100	1	3	8	10	21	Sound; shaft underfilled by 5/8 inch
A-286	2100	2	5	8	9	22	Sound, shaft underfilled by 7/8 inch
16-25-6	2000	2	5	7	12	24	Sound; shaft underfilled by 7/8 inch
Hastelloy C	2200	3	5	10	18	33	Ruptures; shaft underfilled by 1 inch
Hastelloy B	2200	2	4	18	9	31	Ruptures; shaft underfilled by 1-1/4 inch; forging stopped when rupture noticed

(a) Sound refers to forgings that did not rupture.
Underfill dimensions are given as average amounts based on measured heights of hubs.

Friction and Lubrication

Lubricants are employed principally to reduce forging loads and to control or improve the uniformity of metal flow. They are also useful as parting compounds when forging metals with tightly adhering oxides. The following characteristics can be important in the selection of a lubricant:

(1) Insulating qualities. It is usually desirable in forging metals at high temperatures to minimize chilling of the workpiece by the dies.

(2) Corrosiveness. Some otherwise suitable lubricant materials may corrode the die or workpiece. Sulfur compounds can contaminate hot, nickel-rich alloys.

(3) Permanence. Some volatile lubricants and carriers may not wet the dies or may produce noxious or poisonous fumes. Other materials may collect in die cavities and prevent complete filling.

(4) Ease of Removal. This is especially important for corrosive lubricants or for those that affect the appearance of forgings.

In general, the selection of forging lubricants usually represents a compromise among several desirable attributes. Graphite, in various dispersions and preparations, appears to be the most widely used lubricant for forging. Press-forging

experiments indicate that graphite-containing lubricants usually result in friction coefficients ranging from 0.1 to 0.4.[1] Under controlled conditions in forging nonferrous alloys at 850 to 1000 F, a reasonably constant friction coefficient of 0.12 was obtained with a variety of graphite lubricants.[4]

Flat Dies. The pressure required for forging between flat dies depends to a marked extent on friction. The ratio of forging pressure to flow strength also depends on the shape of the workpiece and increases with the ratio of cross-sectional area to height.[2,3,4,5] These effects can be predicted by the formulas listed in Table 2-4, which were developed by Schroeder and Webster.[5]

Figure 2-20 shows a series of curves computed from these formulas. The relationships were found to agree closely with experimental data[3,5,6] obtained at temperatures up to those normally used for pressing aluminum, lead, and magnesium. The effects of variations in disk geometry and in relatively low friction coefficients are pronounced. For example, the chart shows that the lower the friction coefficient, the smaller the effect of the radius/height ratio. For disks with a radius/height ratio of 21, the forging pressure is nine times the flow stress in unlubricated dies. The required pressure can be reduced to five times the flow stress by lubricants producing friction coefficients of 0.10. For such reasons, lubricants for forging in flat dies are ordinarily chosen on the basis of their frictional characteristics.

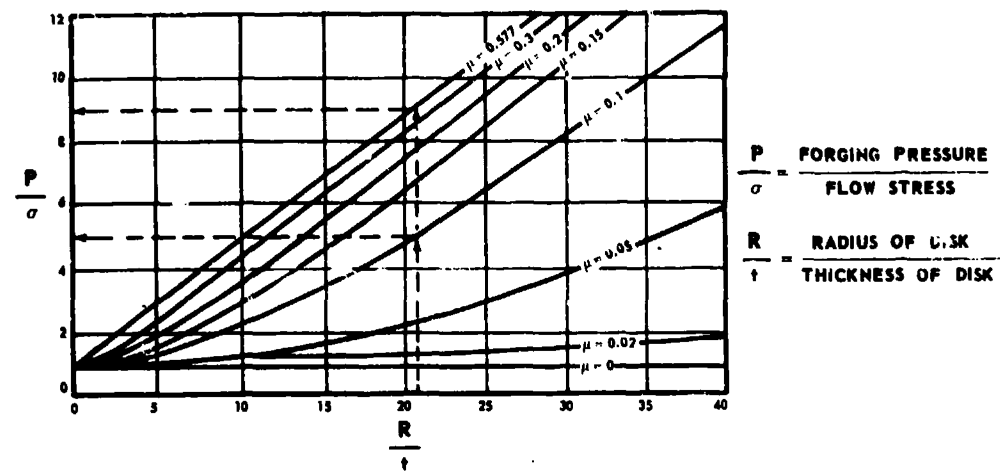


FIGURE 2-20. NONDIMENSIONAL PRESENTATION SHOWING THE COMBINED INFLUENCE OF FRICTION COEFFICIENTS AND DISK GEOMETRY ON THE RATIO OF FORGING PRESSURE TO FLOW STRENGTHS

VALUES CALCULATED USING EQUATIONS IN TABLE 2-4

Shaw and Boulger[6] made an extensive study of forging lubricants for 2014 aluminum. Table 2-5 shows some of their data on frictional characteristics of typical lubricants as evaluated by pressing with flat dies. Depending on the lubricant, a load of 138,000 pounds produced disks ranging from 0.067 inch to 0.101 inch in thickness

TABLE 2-4. FORMULAS RECOMMENDED BY SCHROEDER AND WEBSTER[5] FOR PREDICTING EFFECTS OF SHAPE AND FRICTION ON PRESSURES FOR FORGING CIRCULAR DISKS

Case 1 – When sliding occurs between surfaces of blank and die at all points except the geometric center of blank.

Condition: $\mu << k$ and $\frac{P}{\sigma} < \frac{k}{\mu}$

(friction coefficient less than 0.1 and small $\frac{R}{t}$ ratio)

Solution: $P = m \cdot \sigma$

$$m = \frac{2}{C^2}(e^C - C - 1) \qquad (1)$$

Case 2 – When sliding occurs in an annular zone near the edge but not in the central zone

Condition: $\mu < k$ and $\frac{P}{\sigma} > \frac{k}{\mu}$

(friction coefficient between 0.1 and 0.5)

Solution: $P = m \cdot \sigma$

$$m = \frac{2}{C^2}\left[(D + 1)e^{C-D} - C - 1\right] + \frac{D^2}{C^2}\left(\frac{k}{\mu} + \frac{2k}{3}\frac{r_c}{t}\right) \qquad (2)$$

Case 3 – When sliding does not occur and spreading results from shear strain parallel to the die surface

Condition: $\mu \geq k$

Solution: $P = m \cdot \sigma$

$$m = 1 + \frac{2k}{3}\frac{R}{t} \qquad (3)$$

Nomenclature

P = forging pressure

m = pressure multiplication factor

σ = flow stress of workpiece

μ = friction coefficient

k = constant = 0.577

R = radius of disk

t = thickness of disk

$C = 2\mu\frac{R}{t}$

$D = 2\mu\frac{r_c}{t}$

$\frac{r_c}{t} = \frac{R}{t} - \frac{1}{2\mu}\ln\frac{k}{\mu}$

from standard specimens. The corresponding variations in forging pressure ranged from 23,500 to 36,000 psi. Under the conditions employed, the friction coefficients varied from 0.06 to 0.16. Although a large number of organic and inorganic materials and mixtures were tried, commercial graphite-containing lubricants gave the best results.

TABLE 2-5. EFFECT OF VARIOUS LUBRICANTS IN PRESSING 2014 ALUMINUM ALLOY DISKS BETWEEN FLAT DIES[5]

Billet Size: 1" Diameter x 0.5" High
Billet Temperature: 825 F
Die Temperature: 700 F
Load: 138,000 pounds

Commercial Lubricant[a]	Average Values for Test Series: Final Thickness, inch	Final Diameter, inch	Forging Pressure, psi	Friction Coefficient
Silicone grease	0.103	2.2	36,000	0.16
Flake graphite in oil	0.101	2.2	36,000	0.15
25% flake graphite + 15% MoS_2 + 5% mica in bentone grease	0.095	2.3	33,400	0.13
Colloidal graphite in oil	0.091	2.3	32,000	0.12
Colloidal graphite in water	0.086	2.4	30,000	0.11
30% flake graphite + 5% mica in calcium-base grease	0.067	2.7	23,500	0.06

(a) Commercially available under various trade marks.

Impression Dies. Lubricants suitable for forgings in flat dies do not always give good results in impression dies because the requirements are dissimilar. Low friction coefficients favor lateral metal flow, the main requirement in flat dies. In most impression dies, however, lateral flow must be selectively impeded in order to insure enough vertical flow to fill the cavities. This is one of the principal functions of the narrow, annular, flash openings. Consequently, an ideal lubricant for closed dies would provide minimum friction on vertical surfaces of cavities and maximum friction near the flash openings. These contradictory requirements can sometimes be met by using different lubricants at critical locations.

In principle, lubricant characteristics should influence die design. A particular set of dies cannot be expected to work equally well with lubricating materials having distinctly different characteristics. The practical importance of this interrelationship was demonstrated by Shaw and Boulger[6] in plant tests on lubricants that had performed well in the laboratory. With some die configurations, low-friction lubricants allowed the metal to reach the flash gutters before the die cavities were completely filled. In tests on other forgings, the superior lubricants improved the uniformity of die filling.

The variety of conditions and die designs encountered in commercial operations complicates the task of evaluating forging lubricants. Nevertheless, laboratory studies have provided useful information about lubricating characteristics desirable for forging in impression dies. Therefore, the data in Tables 2-6 and 2-7 are of interest. The results were obtained with different lubricants in forging T-shaped parts from 1-inch-diameter by 1-15/16-inch-long rods in closed dies under a constant punch pressure. The depth of penetration into the die cavity was used to judge the effectiveness of various lubricating materials. It should be recognized that very little lateral flow took place in forging these riblike specimens. Therefore, the ratings of the lubricants are mainly of interest for extrusion-type forgings.

Tables 2-6 and 2-7 show that appreciable variations in die filling were encountered when different lubricants were used under otherwise similar conditions. For the same lubricant, billet pretreatments improved the results obtained in experiments on either magnesium or aluminum. Specifically, etching, drying, and coating the billets with the lubricant before they were heated for forging resulted in better die filling and surface finishes. The effects resemble those of the phosphate treatments given steel blanks to improve lubrication during cold extrusion. Most of the data in Table 2-6 were obtained in dies heated to 700 F, a temperature which gave better results than did lower die temperatures. The benefits of higher die temperatures are now quite widely recognized. It is fairly common practice to forge the AZ80 magnesium alloy in dies heated to 600 F instead of the 500 F temperature mentioned in Table 2-7.

Die Temperature

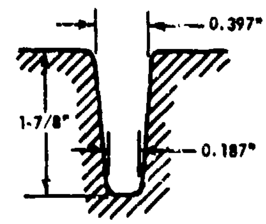
Several investigators have demonstrated that the use of heated dies improves die filling and reduces forging pressures.[6,7,8,9] The benefits result from the fact that hot dies do not chill the surface layers of the workpiece so drastically. Consequently, the effects are more noticeable when dies are heated close to the billet temperatures. This is feasible with metals like aluminum and magnesium but is extremely difficult for titanium alloys, steels, and refractory metals. Temperature changes in the workpiece increase with time as well as with the temperature differential between the die and the billet. Therefore, heated dies offer more advantages when forging with hydraulic presses than with drop hammers and high-velocity machines.

Use of hot dies limits the choice of lubricants and die materials. Most water-base and oil-base forging lubricants are unsatisfactory for die temperatures above 500 F and 800 F, respectively. Ordinary die steels soften when heated above about 800 F but hot-work types (e.g., H-11 and H-12) can be heated to about 1100 F. Superalloys and refractory metals offer promise for dies to be used above 1200 F. The cast Inconel Alloy 713C has been used successfully at die temperatures of 1600 F.[9] The dies were heated with internal resistance elements. High die temperatures are difficult to maintain and control, and the unusual die materials are expensive. Principally for economic reasons, dies are rarely heated above about 1000 F for production operations.

The specific heat and thermal conductivity of the workpiece influence the effects of die temperature on metal flow and die filling. Materials with a relatively low specific heat but good thermal conductivity, like molybdenum, will be chilled rapidly, but the variations in temperature from surface to center will be small. Conversely, metals like titanium with poor conductivity can develop appreciable temperature differentials.

TABLE 2-6. INFLUENCE OF VARIOUS TYPES OF LUBRICANTS ON DEPTH OF FILL IN A T-SHAPED DIE WITH ALUMINUM

Alloy: 2014
Forging Pressure: 46,000 psi
Die temperature: 700 F, unless otherwise noted
Forging temperature: 825 F
Billet size: 1-in. diameter x 1-15/16 in. long

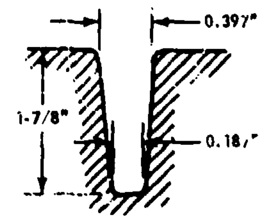


Lubricant[a]	Average Depth of Fill, in	Surface Condition[b]	Coefficient of Friction[c]
	Graphite Suspensions		
Flake graphite in oil			
(500 F dies)	0.78	A	--
(700 F dies)	1.42	A	0.15
Colloidal graphite in oil	1.45	A	0.12
Colloidal graphite in water	1.19	C	0.11
Flake graphite in water	1.25	C	--
Flake graphite + mica in calcium-base grease	1.64	C	0.06
	Molybdenum Disulfide		
MoS_2 in oil	1.28	B	0.10
MoS_2 + mica in calcium-	1.51	B	0.09
base grease			0.09
	Greases		
Soapless grease	0.94	A	0.17
Silicone grease	1.10	B	0.16
Grease-type, cold-forging compound	0.94	A	0.17
	Billets Etched in Caustic, Dried, and Dipped in Suspensions of Graphite		
As coated	1.20	B	--
Graphite in oil	1.87	AA	--
Graphite in water	1.51	A	--

(a) Lubricants were applied by spraying except where noted.
(b) Surface ratings: AA - no scoring
A - slight scoring on radius
B - slight scoring on surface
C - severe scoring on surface.
(c) Determined by disk-forging method.

TABLE 2-7. INFLUENCE OF VARIOUS TYPES OF LUBRICANTS ON DEPTH OF FILL IN A T-SHAPED DIE WITH MAGNESIUM[1]

Alloy: AZ80
Forging Pressure: 46,000 psi
Die temperature: 500 F
Forging temperature: 675 F
Billet size: 1-in. diameter x 1-15/16 in. long



Lubricant[a]	Average Depth of Fill, in.	Surface Condition[b]
Die Lubricants Only		
Flake graphite in oil	1.11	A
Colloidal graphite in oil	0.92	AA
Colloidal graphite in water	1.44	B
Graphite + MoS_2 in water	1.62	AA
Billets Precoated, Dies Sprayed With Oil-Flake Graphite Suspension		
Billets vapor blasted and dipped in aqueous colloidal graphite	1.87	A
Billets degreased and dipped in aqueous colloidal graphite	1.76	C
Billets etched in acetic acid, rinsed, and dipped in aqueous colloidal graphite	1.86	A

(a) Lubricants applied by spraying, except as otherwise noted.
(b) Surface ratings: AA - no scoring
A - slight scoring at radius
B - slight scoring on surface
C - severe scoring on surface.

The relationship between temperature and flow strength, which controls the forging pressure requirements, is another factor that influences the effect of die chilling. Figure 2-21 shows the effects of relatively small temperature changes on the forging pressures of three alloys. It indicates that chilling of the billet by the die would have little or no effect on forging pressures for steel and molybdenum. On the other hand, the effect is pronounced for the titanium alloy and flow becomes concentrated in the hotter, central portions of the forging. When cylinders of the three metals are upset under comparable conditions, titanium exhibits the largest dead-metal zone of undeformed metal. Observations of this kind[10] indicate that the effects of die chilling are more pronounced for:

Titanium alloys	Zirconium alloys
Beryllium	Nickel-base superalloys
Magnesium alloys	

and less marked for:

Steels	Refractory metals
Aluminum alloys	Copper-base alloys

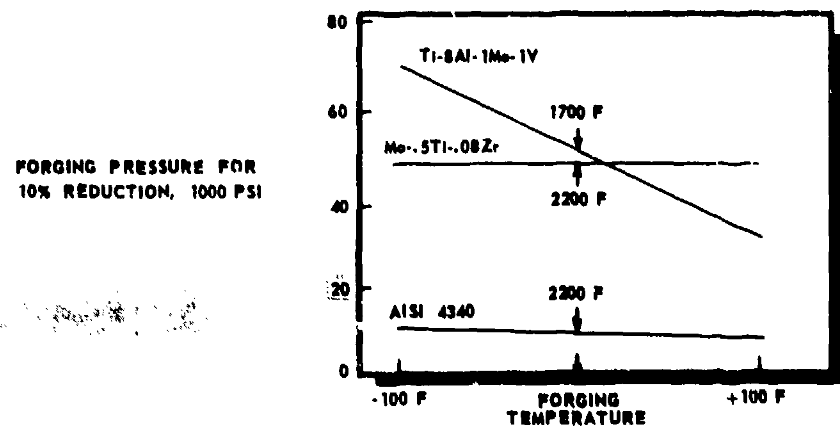


FIGURE 2-21. INFLUENCE OF TEMPERATURE CHANGES ON THE FORGING PRESSURES OF THREE ALLOYS AT THEIR RESPECTIVE FORGING TEMPERATURES[10]

Some materials, such as nickel-base superalloys, exhibit satisfactory forgeability in only a relatively narrow range of temperature. It is possible, therefore, for chilling of the billet by cold dies to cause cracking as well as nonuniform metal flow.

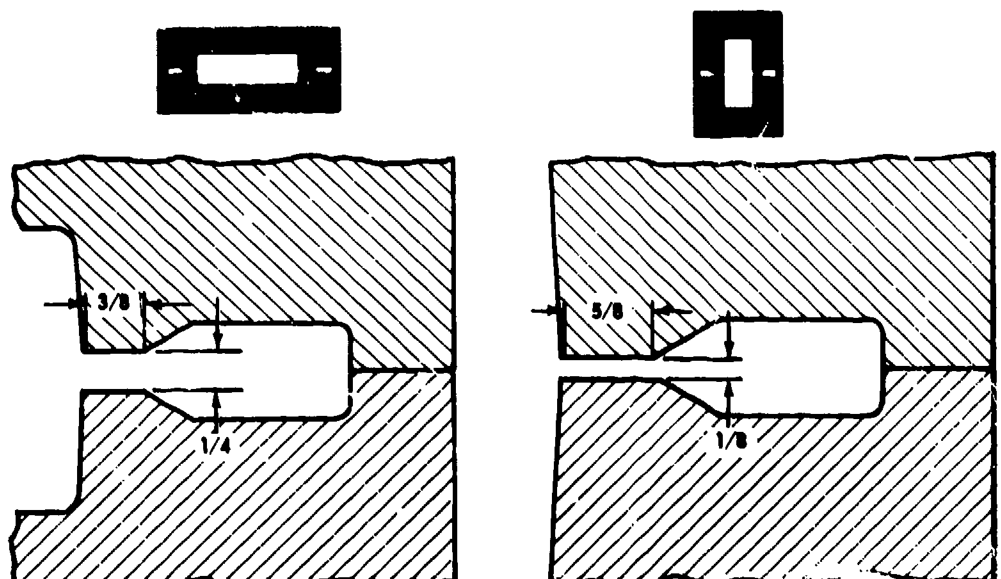
On an experimental basis, some success has been attained in using insulating materials (e.g., glass) to prevent chilling of the surface of the billet. One of the disadvantages of this practice is that the coatings may remain in the die cavities.

Shape and Size Factors

Spherical and blocklike shapes are the easiest to forge in closed dies. Parts with thin and long sections or projections (webs and ribs) are more difficult because

they have more surface area per unit volume. Such variations in shape maximize the effects of friction and temperature changes and, hence, influence the final pressure required to fill die cavities. Neglecting the problems of forging defects, there is a direct relationship between the surface/volume ratio of a forging and the difficulty of producing it. This point is illustrated, for flat disks, in Figure 2-20. Sometimes the difficulties of controlling the direction of metal flow in closed dies also have substantial effects on forging loads.

Vertical projections such as ribs and bosses are appreciably harder to forge than are lateral projections of comparable dimensions. Forging loads are higher because additional lateral constraint is needed to insure complete die filling. Restraint must be provided during the entire time required to form a vertical projection; in a die filled by spreading, this is necessary only when the metal reaches the periphery. Figure 2-22 compares the flash designs required for forging disks in horizontal and in vertical positions. The thinner and wider flash design suggested for the disk forged vertically produces more lateral confinement during the critical period of deformation. Roughly speaking, vertical configurations are characterized by one-third the die life and three times the flash-metal loss of horizontal configurations.



HORIZONTAL DISK
DEFORMATION BY SPREADING

VERTICAL DISK
DEFORMATION BY EXTRUSION

FIGURE 2-22. COMPARISON OF FLASH DESIGNS FOR 10-INCH-DIAMETER X 2-INCH-THICK DISKS FORGED IN HORIZONTAL OR VERTICAL POSITIONS

The ease of forging more complex shapes depends on the relative proportions of vertical and horizontal projections on the part. Figure 2-23 illustrates the fact that relative dimensions of ribs and webs affect the forging operation. Bulky projections on thinner plates have little or no effect. Shapes with tall, slender projections are more difficult to forge than plates or disks. Figure 2-24 is another schematic representation of the effects of shape on forging difficulties. Parts "C" and "D" would not only require higher forging loads but at least one more forging operation than the other to insure die filling.

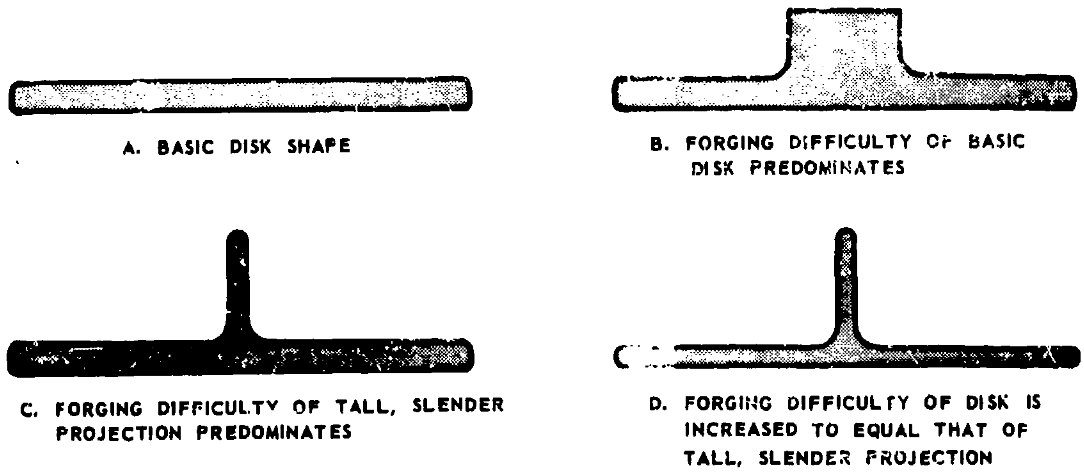
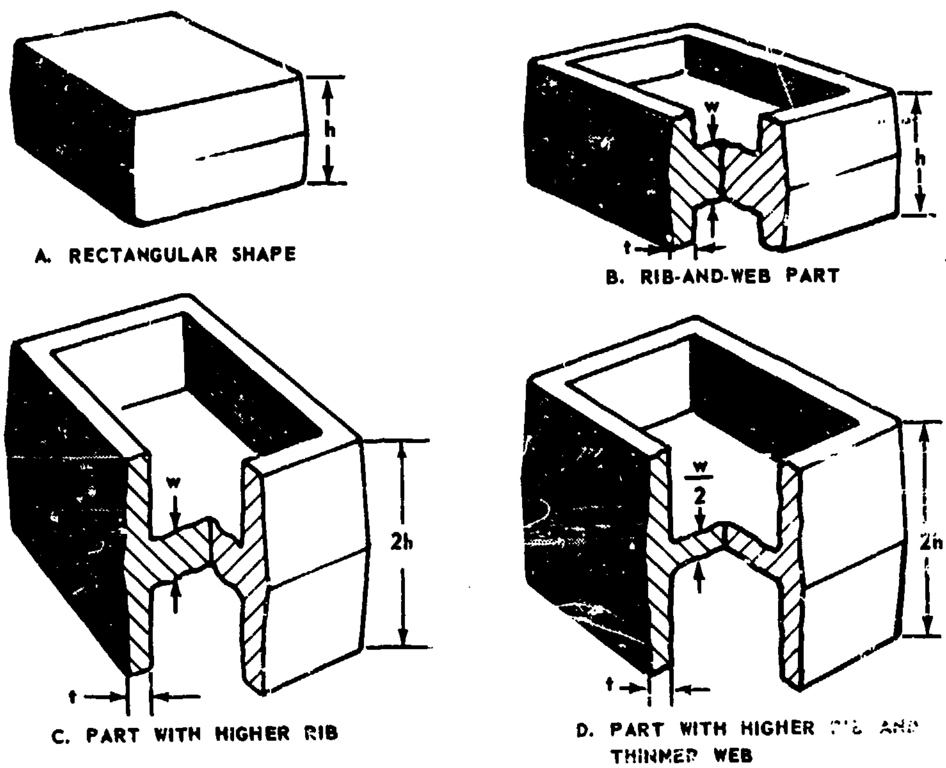


FIGURE 2-23. DISK SHAPE AND THREE MODIFICATIONS THAT ALTER FORGING DIFFICULTY



FORGING DIFFICULTY INCREASES FROM A TO D

FIGURE 2-24. RECTANGULAR SHAPE AND THREE MODIFICATIONS SHOWING INCREASING FORGING DIFFICULTY WITH INCREASING RIB HEIGHT AND DECREASING WEB THICKNESS

The increase in load during the progress of forging in impression dies is shown schematically in Figure 2-25. Loads are comparatively low until the more difficult details are partly filled and the metal reaches the flash opening. This stage corresponds to Point P_1 in the sketch. For successful forging, two conditions must be fulfilled when this point is reached:

(1) A sufficient volume of metal must be trapped within the confines of the die to fill the remaining cavities.

(2) The difficulty of extruding metal through the narrowing gap of the flash opening must exceed that of filling the most difficult detail in the die.

As the dies continue to close, the loads increase sharply to Point P_2, the stage where the die cavity is filled completely. This would correspond to the maximum forging load if designs were figured that closely and no flash was formed. In that respect, the most difficult detail determines the minimum load for producing a satisfactory forging. Usually, however, flash is necessary and the work done in extruding metal through the narrow flash opening causes the load to increase from P_2 to P_3. Therefore, the dimensions of the flash determine the load required for closing the die or the maximum load developed during forging.

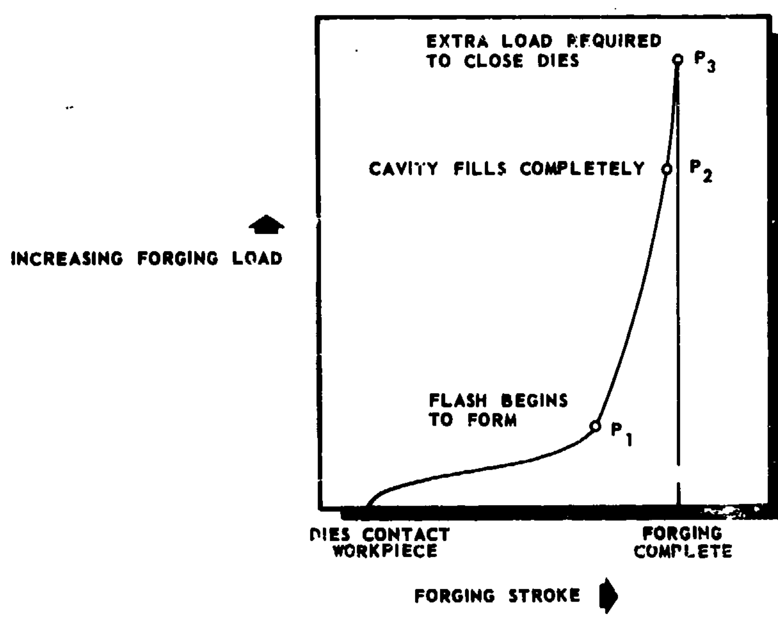
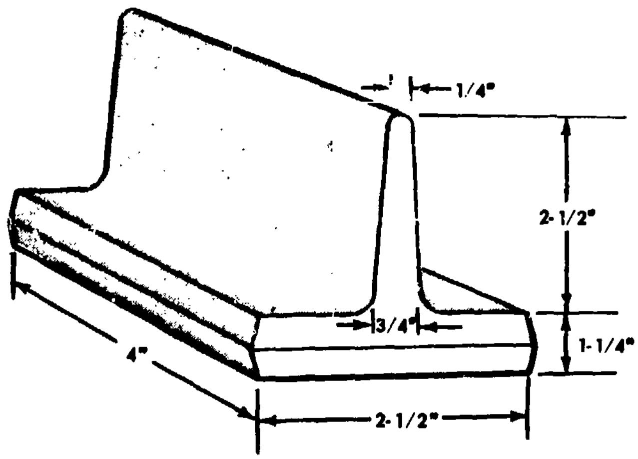


FIGURE 2-25. TYPICAL FORGING-LOAD CURVE FOR CLOSED-DIE FORGING, SHOWING THREE DISTINCT STAGES

MacDonald[11] has shown that the load required for deforming the flash is approximately the same as that for forging a thin slab of similar dimensions. Therefore the die-closing load can be estimated from the thickness and projected area of the forging and flash and from the flow stress of the material by using the relationships illustrated in Table 2-4 and Figure 2-20. For example, experience shows that a flash opening 0.50 inch wide and 0.10 inch thick is suitable[12] for a ribbed component of the type shown in Figure 2-26. This example indicates that forging such a part from a material



Flash dimensions: 0.50 in. wide, 0.10 in. thick

Projected area: (4 + 1/2 + 1/2)(2-1/2 + 1/2 + 1/2) = 17-1/2 sq. in.

Radius of disk of equal area = $\sqrt{\dfrac{17.5}{\pi}}$ = 2.36

Radius/thickness $\left(\dfrac{R}{h}\right) = \dfrac{2.36}{0.1} = 23.6$

Pressure multiplication factor,
from Equation (3) of Table 2-4 = 10.2

Forging load for material with flow stress of 6,000 psi:
F = 17.5 x 6,000 x 10.2 = 1,071,000 lbs ≅ 535 tons

FIGURE 2-26. ILLUSTRATION OF METHOD OF ESTIMATING LOAD REQUIRED FOR FORGING A HYPOTHETICAL SHAPE FROM FLASH DIMENSIONS

with a flow stress of 6,000 psi should require a load of about 535 tons. This value is a close approximation of the load required for forging the 7075 aluminum alloy to this shape at 700 F. Similarly, the forging load estimated by this method checks the one computed by Dietrich and Ansel[12] to exemplify their more complicated analysis.

REFERENCES

(1) Peterson, M. B., and Johnson, R. L., "Friction Studies of Graphite and Mixtures of Graphite With Several Metallic Oxides and Salts at Temperatures to 1000 F". NACA Technical Note No. 3657 (February, 1956).

(2) Siebel, E. E., A series of articles on "The Plastic Forming of Metals", Steel, Vols 93 and 94 (October 16, 1933 through May 7, 1934).

(3) McDonald, A. G., Kobayashi, S., and Thomsen, E. G., "Some Problems of Press Forging Lead and Aluminum", ASME Trans.: Series B, J. of Eng. for Ind., pp. 246-252 (August, 1960).

(4) Tholander, E., "Forging Research In Sweden", Metal Treatment and Drop Forging, V. 28, pp. 89-94 (March, 1961).

(5) Schroeder, W., and Webster, D., "Press Forging Thin Sections: Effect of Friction, Area, and Thickness On Pressure Required", Trans. ASME: J. App. Mech., Vol 71, pp. 289-294 (September, 1949).

(6) Shaw, H. L., Boulger, F. W., and Lorig, C. H., "Development of Die Lubricants For Forging and Extruding Ferrous and Non-Ferrous Materials", Summary Report on Contract AF-33(600)-26272, Battelle Memorial Institute (October 21, 1955).

(7) Proffitt, C. E., and Thomas, H. C., "Close-Tolerance Steel Forging Development", Boeing Airplane Company, Final Technical Engineering Report on Contract AF 33(600)-36659 (April, 1962).

(8) Knauschner, A., "Influence of Die Temperature and Die Lubrication on the Rising of Steel in the Die", Fortigungstech u. Betrieb, 10, 139-145 (March, 1960).

(9) Gouwen, P. R., "Heated Dies Produce Better Forgings", Am. Machinist, V. 103, pp. 128-131 (October, 1959).

(10) Henning, H. J., Spretnak, J. W., Sabroff, A. M., Boulger, F. W., "A Study of Forging Variables", ASD Interim Report 61-7-876(3) on Contract AF 33(600)-42963, Battelle Memorial Institute (May, 1962).

(11) MacDonald, A. G., Kobayashi, S., and Thomsen, E. G., "Approximate Solutions to a Problem of Press Forging", Jrl of Engineering for Industry, pp. 21[illegible]-227 (August, 1959).

(12) Dietrich, R. L., and Ansel, G., "Calculation of Press Forging Pressures and Applications to Magnesium Forgings", Trans. ASM, V. 38, pp. 709-728 (1947).

CHAPTER 3

PRINCIPLES OF DIE-FORGING DESIGN

DESIGN FACTORS

Forging design is much like the design for other metalworking processes – it is influenced by the nature of the metal being processed and the capabilities and limitations of the available forging equipment and tools. The more plastic materials can be forged in the greatest variety of shapes and designs. But it must be remembered that there are design limitations even for the most highly plastic alloys. For alloys with less plasticity or forgeability, the design limits are often quite narrow and a knowledge of these is essential to the design engineer in considering the use of a forged part for a structural component. This chapter describes the important fundamentals of forging design and the relationships between design, forging materials, forging equipment, and tolerances attainable on designs.

Parting-Line Location

Given a certain shape intended for forging, the first step in forging design is to establish the location and shape of the parting line (sometimes called "flash line" or "split line"). This is the line along which the dies separate. The parting line may be straight or irregular, depending on the geometry of the final part. This first decision influences other design factors such as die design and construction, grain flow, and trimming procedure. Principles for locating the parting line vary, depending on the type of available forging equipment and, to some extent, on the alloy being forged. Figure 3-1 illustrates a variety of simple shapes showing undesirable and preferred parting-line locations. Reasons for the preferred locations are:

Case 1. The preferred choice avoids deep impressions that might otherwise promote die breakage.

Cases 2 and 3. The preferred choices avoid side thrust which would cause the dies to shift sideways.

Case 4. Preference is based on grain-flow considerations. The "satisfactory" location provides the least expensive method of parting, since only one impression die is needed. The preferred location, however, produces the most desirable grain-flow pattern.

Case 5. The choice in this case is also based on grain-flow considerations. However, the "desirable" location usually introduces manufacturing problems and is used only when grain flow is an extremely critical factor in design. This, in turn, depends on the directional properties and cleanliness of the metal being forged.

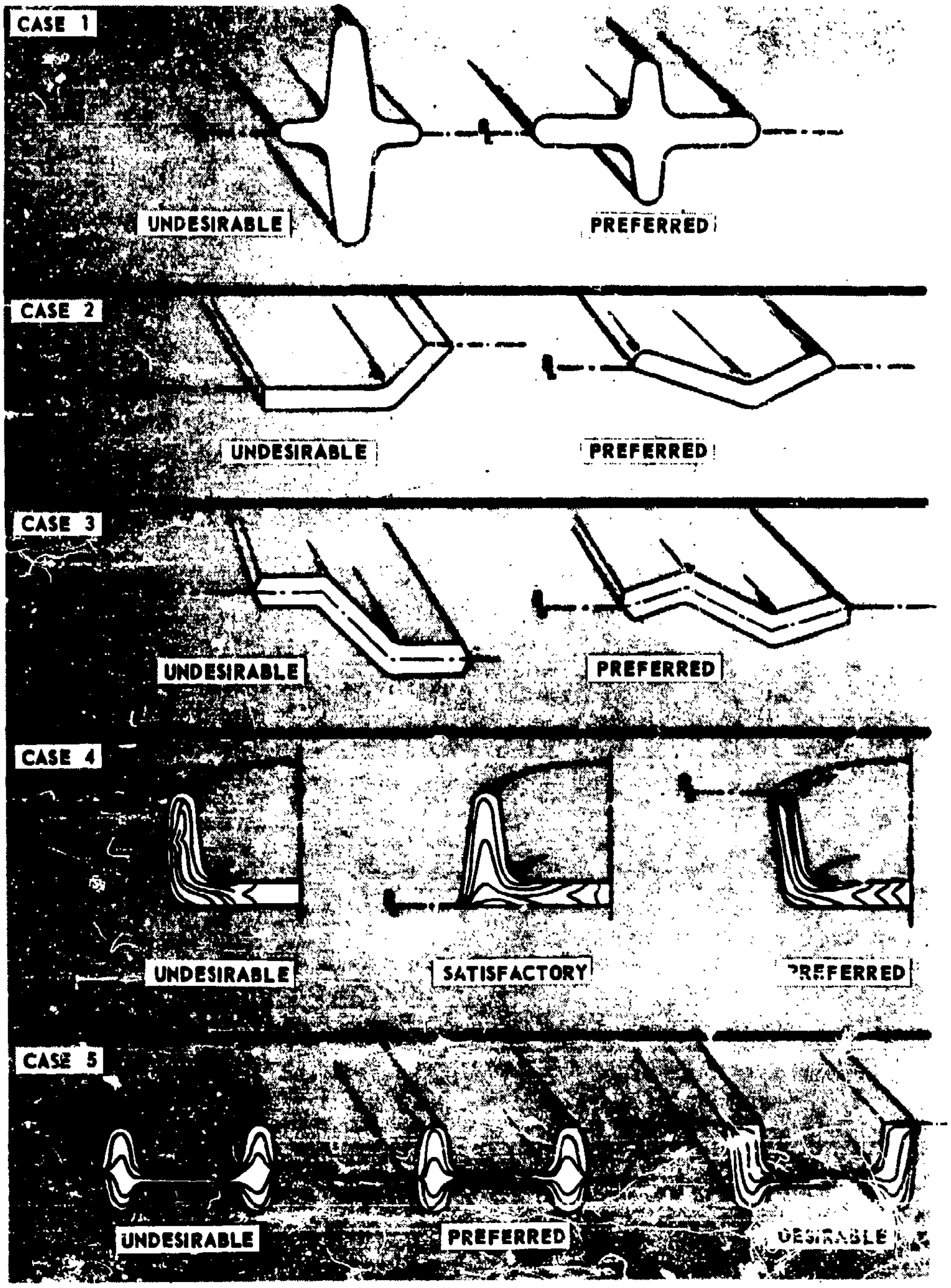


FIGURE 3-1. SEVERAL TYPICAL FORGING SHAPES ILLUSTRATING UNDESIRABLE AND PREFERRED PARTING-LINE LOCATIONS

Methods for locating the parting line on complex shapes are not as straightforward as these examples would indicate. However, they illustrate most of the fundamentals involved.

Finish Allowance

Finish allowance (sometimes called cleanup allowance, forging envelope, or machining allowance) is the amount of excess metal surrounding the intended final shape. The finish allowance depends, to a great extent, on the oxidation behavior of the metal being forged. Since aluminum and magnesium alloys do not oxidize appreciably at forging temperature, they are often forged without finish allowances. Most other metals, however, are subject to oxidation, contamination, or decarburization at their respective forging temperatures. Hence, a finish allowance is usually provided so that affected metal can be removed by machining. The minimum finish allowance usually depends on how much surface material has to be removed. Also, certain additional finish allowances are generally provided to compensate for other possible local surface defects. Finish allowances usually increase with increasing forging size because of longer heating times, added operations, and a greater chance for nicks, dents, and other defects to occur during handling.

Figure 3-2 indicates practical finish allowances recommended for several forging-alloy systems. Forgings with smaller finish allowances are obtainable but usually at added cost.

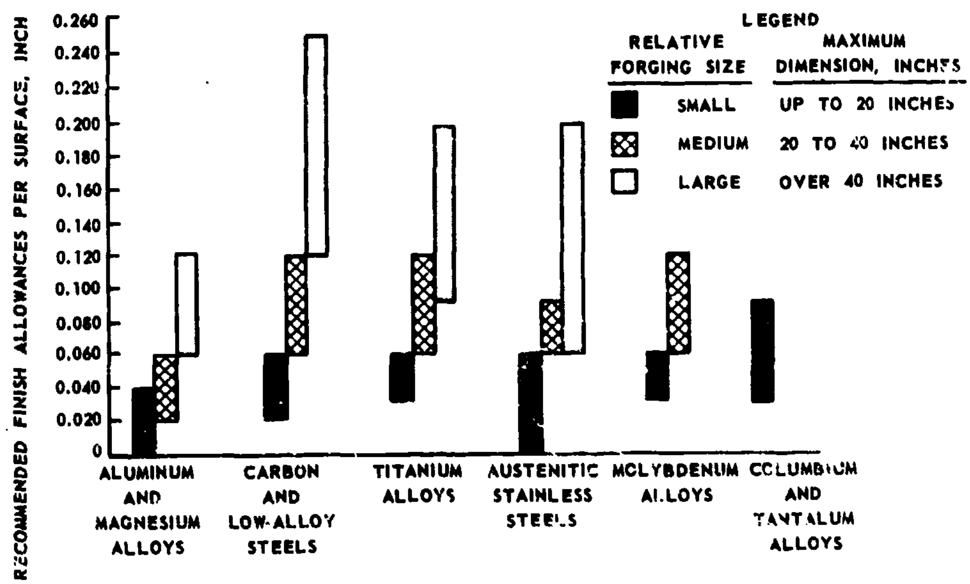
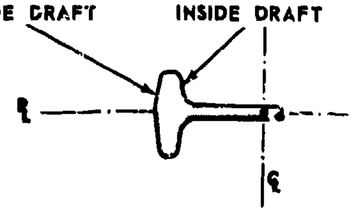


FIGURE 3-2. RECOMMENDED FINISH ALLOWANCES FOR DIE FORGINGS OF VARIOUS ALLOY SYSTEMS[1]

Forging Draft

Axial projections on a forging are usually tapered so that the forging can be easily removed from the die cavity. This taper is called draft. A typical draft for a disk-shaped forging is shown below. The most common draft angles are between 5 and 7



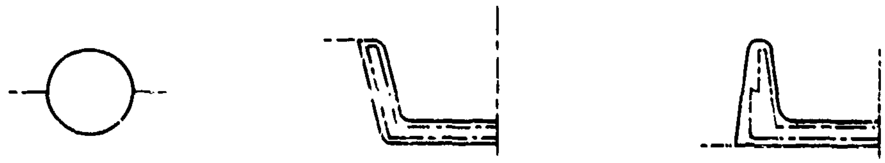
degrees. Smaller draft angles can be used but will usually result in increased production difficulty. For steel forgings, it is common to apply a smaller draft angle on the outside surface than on the inside, because the outside surface will shrink away from the die during cooling and permit removal of the forging. Forging designs with zero draft require dies with special knockouts.

Natural Draft. After the parting line has been established and machining allowances provided, a shape may have what is called natural draft. Figure 3-3 shows five shapes that exhibit natural draft on all or some of their surfaces. The sketches also show how parting-line changes can be made to take advantage of natural draft.

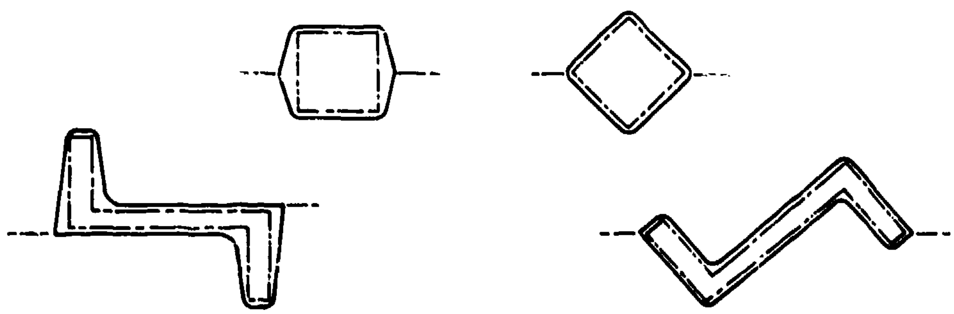
Matching Draft. When unsymmetrical ribs and sidewalls meet at the parting line, it is standard practice to provide greater draft on the shallower die to make the forging surfaces meet at the parting line. This is called matching draft, and is illustrated in Figure 3-4. It is sometimes necessary to use the principle of matching drafts for surfaces other than the outside surfaces of a forging. The lower sketch in Figure 3-4 indicates the desirability of matching the drafts of centrally located ribs that are slightly offset. This is done frequently to provide for the most favorable grain flow in the rib.

Selecting the Draft Angle. It is difficult to apply hard and fast rules for selecting draft angles appropriate to individual forging designs. Seven degrees is the most commonly used draft angle. To reduce machining requirements, however, it is often desirable to seek the absolute minimum draft. This is particularly true for tall profiles. Differences in equipment capability influence the selection of draft. For example, it is undesirable to use knockout equipment in drop hammers because of the risk of die breakage and attendant safety problems. For this reason, forgings with zero-degree draft are seldom considered for drop-hammer forging. On the other hand, forgings with zero draft can be produced on hydraulic and mechanical presses and other equipment more readily adapted to knockout arrangements.

The metal being forged also has a significant influence on the minimum practical draft angles. Metals heated to temperatures far above that of the dies tend to shrink

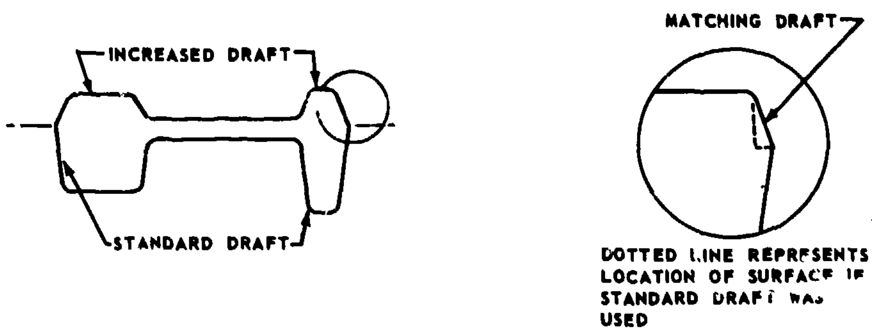


A. NATURAL DRAFT INHERENT IN PART DESIGN

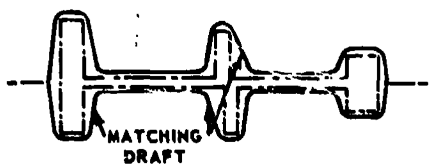


B. NATURAL DRAFT PROVIDED BY CHANGING PARTING LINE

FIGURE 3-3. SEVERAL TYPICAL SHAPES EXHIBITING NATURAL DRAFT



A. UNSYMMETRICAL SIDEWALLS AND RIBS



B. UNSYMMETRICAL AND OFFSET RIBS

FIGURE 3-4. EXAMPLES OF MATCHING DRAFT

away from the die cavity and permit easy removal of forgings. This is particularly true if the metal forms an oxide that serves as a parting agent between the forging and dies. Alloys of titanium and nickel tend to seize or gall in dies of shallow draft because they do not develop a thick oxide during heating, such as the scale on steels. In at least one reported case, a titanium forging actually bonded or welded to the forging die, thus ruining it. Such alloys require greater draft angles than do steels. Guide lines for selecting minimum draft angles for different alloys may be summarized as follows:

(1) Zero and 1-degree draft angles may be used on aluminum and magnesium forgings of the extrusion type. Back-extruded cylinders and shafts are frequently designed with a 1-degree draft. The operations are usually performed either on presses with coaxial rams or by split-die techniques. Complex rib-and-web parts of aluminum and magnesium have been forged with zero draft, providing the ribs are formed by the extrusion method.

(2) A 3- to 5-degree draft angle is suitable for most forgings of carbon, low-alloy, and stainless steels, and for some of the nickel-base alloys. Knockout equipment is desirable but not essential.

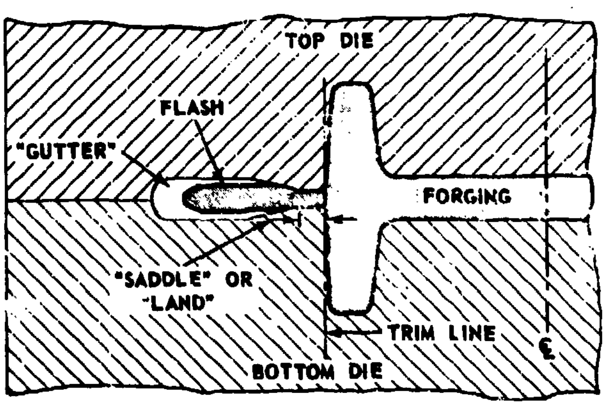
(3) A 5-degree draft angle is generally considered the minimum for titanium alloys because shallower drafts often lead to seizing and galling problems.

(4) A 7-degree or greater draft angle is generally required for forgings of alloys requiring extreme pressures such as refractory metals, the nickel-base superalloys, and the hot-cold worked austenitic stainless steels (e.g., 19-9DL, 16-25-6).

It should be recognized that forging production costs normally increase as the draft angle is reduced below about 7 degrees. Shallower draft angles can be used, however, in special forging equipment such as the high-energy-rate machines. Even titanium and some of the nickel-base alloys have been forged with draft angles of 1 degree or less in such machines.

Flash

Conventional forging dies are designed with a cavity for flash, as shown in the following sketch of a completed forging.



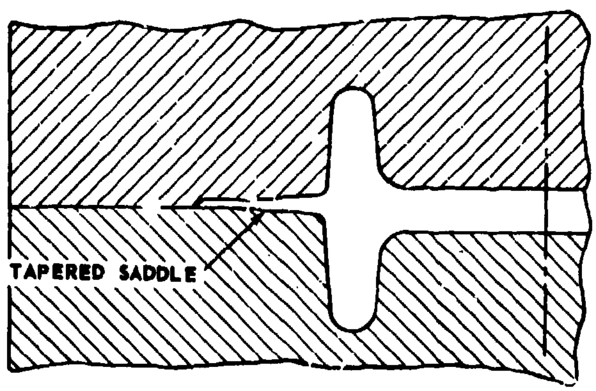
Important dimensions in the flash region of a die are the saddle width and flash thickness. Increasing constraint is provided to the metal being forged by increasing the saddle width and/or by reducing the flash thickness. Restricting the formation of flash by such changes in the die design, however, result in higher forging-load requirements.

The influence of flash geometry on forging-load requirements was determined for a typical magnesium impression die forging of the AZ61 alloy by Dietrich and Ansel.[2] For variations in the saddle-width and flash-thickness dimensions on dies for the same forged part, the forging loads required were:

Flash Dimensions, inch		Forging Load,
Saddle Width	Thickness	tons
0.15	0.046	300
0.10	0.025	385
0.20	0.025	550

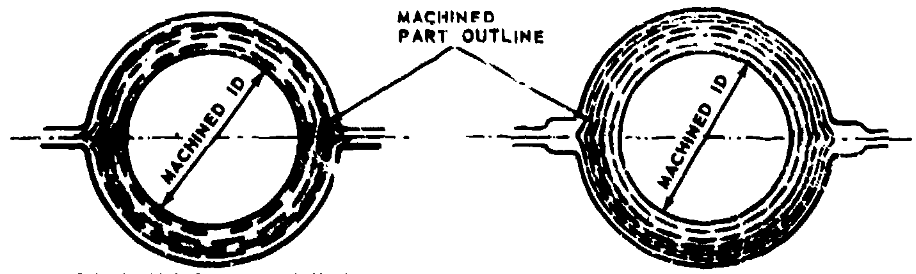
It can be seen that relatively small changes in the flash dimensions can result in appreciable changes in the forging-pressure requirements.

Depending on the size of the forging, flash-metal losses may vary between 5 and 15 per cent. Losses up to 30 per cent have been reported, however, for tall, narrow forgings and for blades. To minimize flash-metal losses, a "tapered saddle" is sometimes used, as shown in the sketch below:



Tapered designs provide considerably more constraint to metal flow than do parallel saddle designs, and thus reduce flash-metal losses. The greater forging pressures required, however, impose greater stresses on forging dies. A tapered-saddle design is generally used when the material savings justify the use of larger forging equipment. To a degree, therefore, the choice of flash design becomes a matter of economics.

At times the outward flow of metal at the parting line assumes added importance in forging design. High-strength-steel cylindrical landing-gear forgings, for example, often exhibit abruptly changing grain-flow patterns in the parting-line region and, as a result, have lower transverse ductility in this region (this is called the "flash-line effect"). A method for minimizing this effect involves the use of a beaded flash line, as illustrated in Figure 3-5, to shift the undesirable grain-flow pattern away from the finished part outline. Figure 3-6 shows an example of a large, high-strength steel die forging with such a bead at the parting line.



a. Cylinder With Conventional Flash Design Exhibiting Interrupted Grain Flow in the Parting-Line Region

b. Cylinder With Bead At The Parting Line Showing Comparatively Smooth Grain Flow In The Parting-Line Region

FIGURE 3-5. METHOD OF IMPROVING GRAIN-FLOW PATTERN IN CYLINDRICAL FORGINGS BY USING A BEAD AT THE PARTING LINE

Fillet and Corner Radii

One of the most important factors in the design of die forgings is the proper selection of fillet and corner radii. Metal flowing in a die containing cavities of various depths will not undergo abrupt changes in the flow direction, and can lead to a condition called "flow-through". This effect is illustrated in Figure 3-7, which compares the metal flow in impression dies designed for large and small fillet radii. A lap or cold shut can form as a direct result of flow-through if the fillet radius is too small.

Forgings made in a series of dies with varying fillets demonstrate the minimum fillet for a particular change in die-cavity depth; this has been determined for a number of materials by the forging industry. Figure 3-8 illustrates how the minimum fillet radius varies with increasing rib height for forgings of average proportions. It should be noted that the fillets given in the upper curve of this illustration are for a single-impression die with the rib near the edge of the impression. The lower curve depicts smaller fillets that can be obtained by using successive forging dies. An acceptable rule of thumb regarding fillets on finished forgings containing ribs is that, to achieve smaller fillets, one additional forging die is required to reduce the fillet radius by 50 per cent. Thus, a forging containing a 3-inch-high rib would require one die for a 1-inch fillet, two successive dies for a 1/2-inch fillet, and three successive dies for a 1/4-inch fillet. It is usually possible to use smaller fillets adjacent to centrally located ribs and bosses. As a general rule, such fillet radii can be about 25 per cent smaller than for ribs near the edge of the impressions.

The choice of corner radii is not as critical as it is for fillets. However, small corner radii, e.g., 1/16 inch, generally increase the chances for die failure and are more difficult to fill. As shown in Figure 3-9, large corner radii are preferred for bosses and ribs, and a full radius is considered optimum for ribs. Minimum recommended corner radii for bosses and ribs are indicated in Figure 3-10 for aluminum-alloy forgings.

Because of differences in forging characteristics, the recommendations for fillets and corner radii differ among materials. For example, where a 1/16-inch corner radius

FIGURE 3-6. LARGE AIRCRAFT LANDING-GEAR FORGING DESIGNED WITH A BEAD AT THE PARTING LINE TO MINIMIZE FLASH-LINE EFFECT

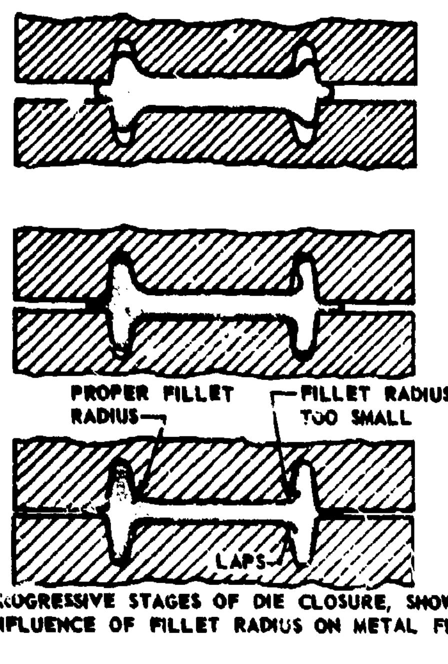


FIGURE 3-7. PROGRESSIVE STAGES OF DIE CLOSURE, SHOWING THE INFLUENCE OF FILLET RADIUS ON METAL FLOW

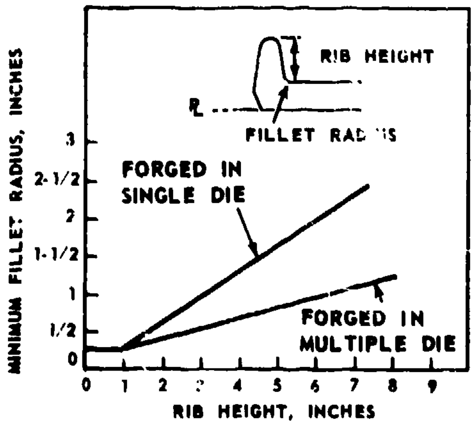


FIGURE 3-8.

INFLUENCE OF RIB HEIGHT ON THE MINIMUM FILLET RADIUS FOR STEEL AND ALUMINUM FORGINGS OF AVERAGE PROPORTIONS(1,3,4)

Data shown are for ribs near outer edge of forging; minimum radii for centrally located ribs may be about 25 per cent smaller.

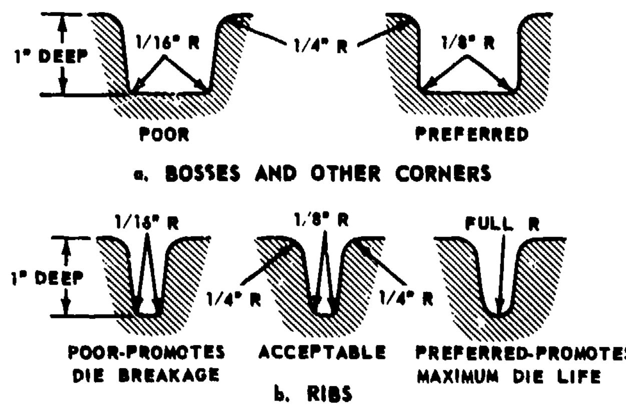


FIGURE 3-9.

OPTIONAL DESIGNS OF CORNER RADII FOR ALUMINUM AND STEEL FORGINGS, SHOWING ORDER OF PREFERENCE FOR 1-INCH-HIGH BOSSES AND RIBS(1,3,4)

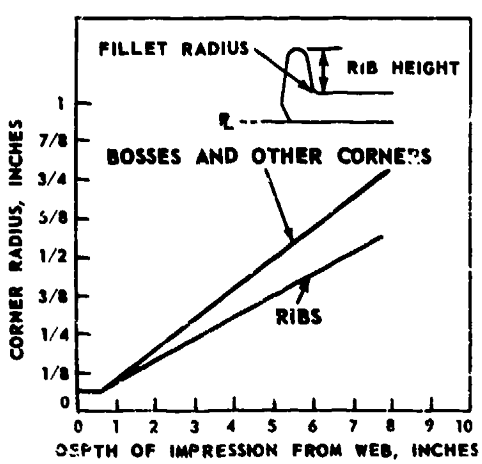


FIGURE 3-10.

INFLUENCE OF IMPRESSION DEPTH ON THE CORNER RADII OF RIBS, BOSSES, AND OTHER EDGES FOR STEEL AND ALUMINUM FORGINGS OF AVERAGE PROPORTIONS(1,3,4)

is used for aluminum forgings, a 1/8-inch corner radius would be used for titanium-alloy forgings of comparable shape and size. Table 3-1 summarizes the fillet and corner radii suggested by a number of forging companies for each of several alloys.

Another guide to the selection of minimum radii for fillets and corners or shallower, less critical projections than ribs or bosses is based on the forging weight. Table 3-2 presents some recommended values for fillet and corner radii based on forging weight. Except for very small forgings, the fillet radii are normally twice the recommended corner radii. For the range of radii given in Table 3-2, the smaller radii are for generally flat, shallow impressions.

It is important to maintain corner radii as consistent as possible for any given shape. This practice minimizes the need for numerous machine-tool changes during die sinking and reduces die costs.

Minimum Section Size

One of the most important methods for reducing forging weight is to avoid excess metal on tall ribs or large webs where small reductions in thickness account for sizable weight reductions. However, there are practical limits on the minimum section thickness because many alloys have narrow forging-temperature ranges or because they require high forging pressures. Thin, flat sections require considerably greater forging pressure than bulkier shapes because they are exposed to greater frictional forces per unit volume, and also because they cool more rapidly. These factors impose appreciably higher stresses on dies and forging equipment.

Another limit on section size is imposed on webs between ribs or other projections. Figure 3-11 shows the sequences in forging two rib-and-web shapes of identical design, except for the web thickness. In forming the ribs, metal normally flows toward the center, causing the web to increase in thickness. If the web in the blocked forging is too thin it will buckle, causing laps in the web. Thus, there are limits imposed on web thickness that depend on the geometry of the forging. The minimum web thickness producible in forgings completely surrounded by ribs is approximately one-eighth the web width (a web thickness-to-width ratio of 1:8). This is by no means a rigid rule. By using carefully designed preliminary dies, forging companies have been able to produce sound, finished forgings with 1/8-inch-thick webs between ribs 3 inches apart. This represents a web thickness-to-width ratio approaching 1:24. As a general rule, however, this is accomplished only by using a series of at least three separate die sets; preblock, block, and finish.

The thickness limits on thin ribs are imposed in much the same manner as those for webs. They are influenced by the metal being forged and by the forging geometry. As described earlier, the choice of fillet radii and the location of the rib play important roles in forging a sound, vertical rib. The location of the parting line also influences the possible rib geometry. These factors are considered in the following discussion of four basic types of ribs, which are shown in Figure 3-12:

Type 1. Centrally located ribs are formed by extruding metal from the body of a forging. Geometric limits are imposed on this type of rib by the volume of metal beneath the rib and by the forging-pressure requirements. If the volume of metal required for filling this type of rib is not available in the body of the forging,

TABLE 3-1. FILLET AND CORNER RADII SUGGESTED FOR FORGINGS OF SEVERAL ALLOYS WITH 1-INCH-HIGH RIBS[1]

Alloy	Fillet Radius, inch		Corner Radius, inch	
	Preferred	Minimum	Preferred	Minimum
2014	1/4	3/16	1/8	1/16
AISI 4340	3/8-1/2	1/4	1/8	1/16
H-11	3/8-1/2	1/4	3/16	1/16
17-7PH	1/4-1/2	3/16	3/16	3/32
A-286	1/2-3/4	1/4-3/8	1/4	1/8
Ti-6Al-4V	1/2-5/8	3/8	1/4	1/8
Unalloyed Mo	1/2	--	Full radius up to 1/2 in.	3/8

TABLE 3-2. GENERAL RECOMMENDATIONS FOR MINIMUM FILLET AND CORNER RADII OF STEEL FORGINGS ON A WEIGHT BASIS[1,4]

(Excluding Ribs and Bosses)

Forging Weight, lb	Fillet Radius, inch	Corner Radius, inch
1	3/64-1/8	[illegible]/64-1/8
2	1/16-1/8	1/16-1/8
5	1/8-1/4	1/8
10	1/8-1/4	3/32-1/8
30	1/4-1/2	1/8-1/4
100	1/2	1/4

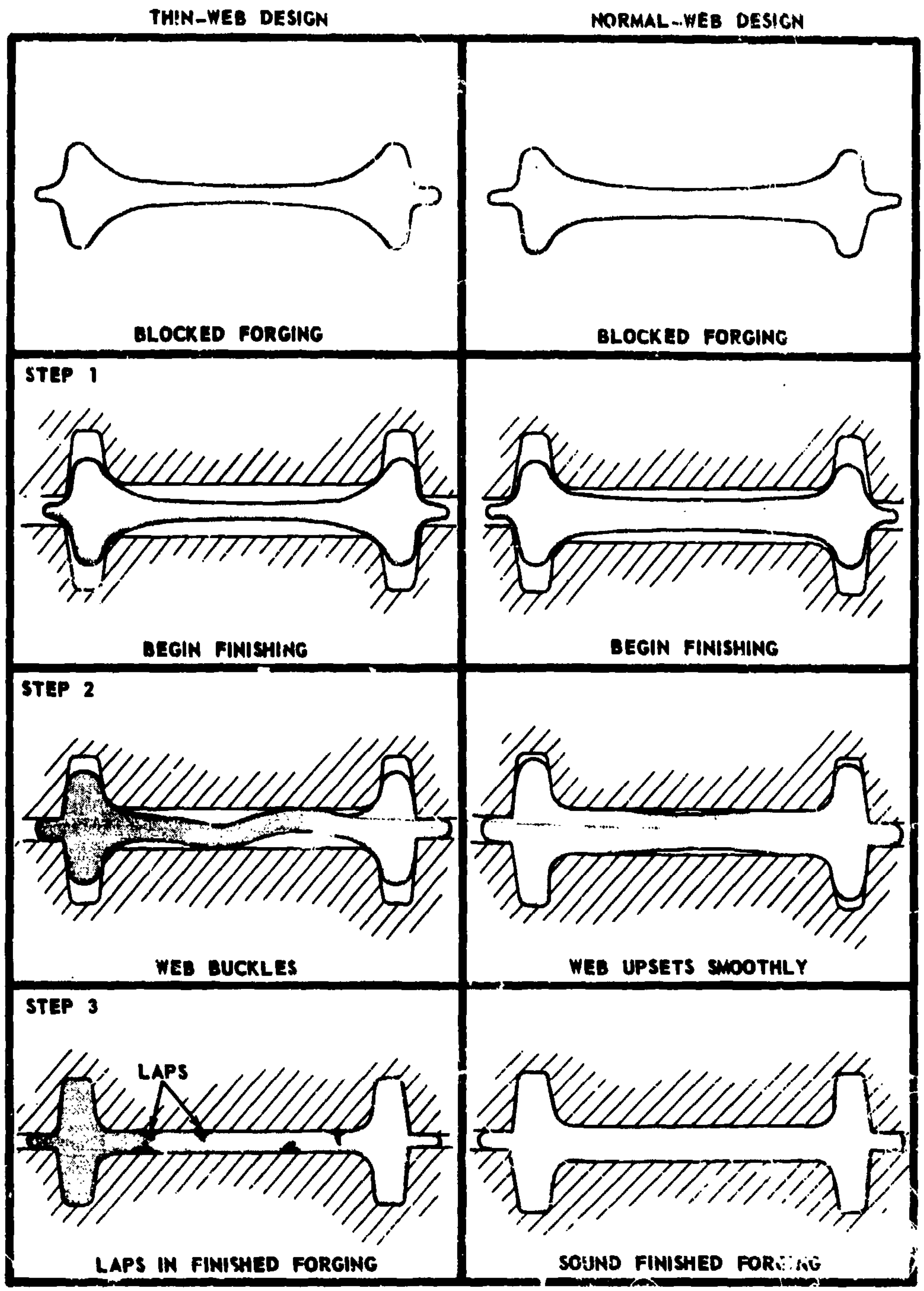


FIGURE 3-11. TYPICAL SEQUENCES IN THE DIE FORGING OF A RIB-AND-WEB PART, SHOWING HOW DEFECTS CAN OCCUR IF WEB IS TOO THIN

a defect similar to "extrusion pipe" may form on the opposite side. As a general rule, the web thickness should be equal to or greater than the rib thickness to avoid defects. In marginal cases, it is sometimes helpful to provide a short rib on the opposite side of the web, as shown in the second sketch. Thicker ribs are needed when forging metals requiring unusually high forging pressures.

Type 2. Ribs at the edges of forgings with the parting line at the top are formed by back extrusion. Such ribs are not subject to forging or extrusion-type defects; therefore, section thicknesses are limited only by material considerations. Ribs thinner than the other three types are possible with this design.

Type 3. Ribs at the edge of forgings with the parting line at the base are also formed by an extrusion type of flow, but are not subject to extrusion-type defects. Section-thickness limits are similar to those for the Type 1 ribs, except that the possibility of flow-through defects limits the rib height to a greater extent.

Type 4. Ribs at the edge of a forging with a central web are the most difficult to forge. Not only are these ribs subject to flow-through defects but they require almost double the volume of metal required for the Type 3 ribs of comparable dimensions. Such rib designs almost always require preliminary tooling to gather the volume of metal necessary for filling the final die. Minimum thicknesses for the Type 4 ribs are generally larger than are those for the other three types.

There are no hard and fast rules that apply to the dimensions of ribs. In general, the rib height should not exceed eight times the rib width. Most forging companies prefer to work with rib height-to-width ratios between 4:1 and 6:1. Table 3-3 summarizes, for several alloys, the suggested minimum section-size limits for ribs and webs with conventional fillet and corner radii and draft angles. The most important fact brought out by this table is that the minimum section sizes normally increase as the forging-pressure requirements increase.

FORGING TOLERANCES

There are practical limitations on dimensions and other characteristics of forged parts, which vary according to the part shape and size. These limitations are influenced by the type of forging equipment, unavoidable variations in forging operations, and the properties of the alloy being forged.

Length and Width. Length and width tolerances allow for variations in dimensions measured in a plane parallel to the parting line of the dies. These tolerances include allowances for shrinkage, die-sinking, and die-polishing variations.

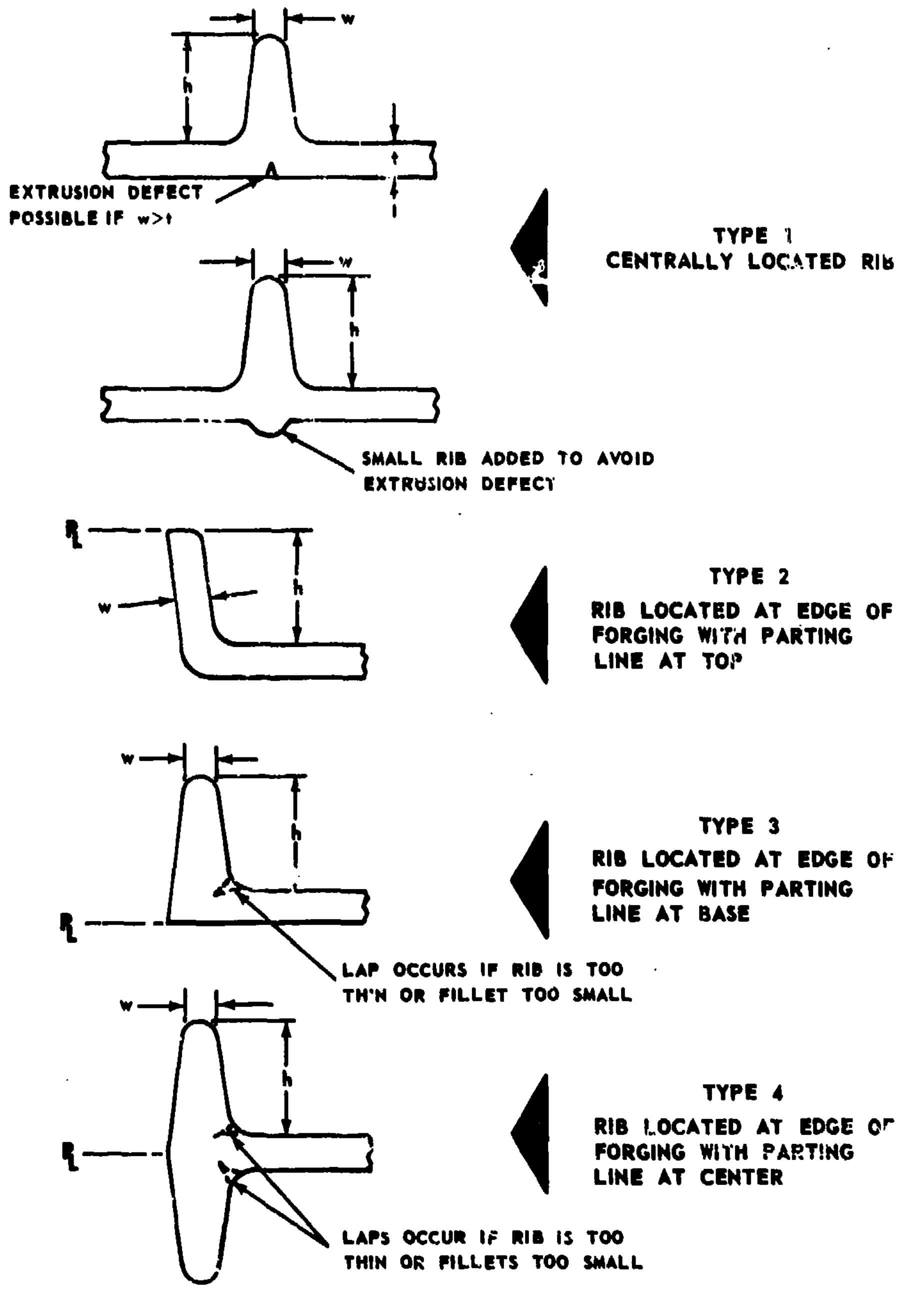


FIGURE 3-12. FOUR BASIC TYPES OF RIB DESIGNS

TABLE 3-3. SUGGESTED MINIMUM SECTION THICKNESSES FOR RIB-AND-WEB FORGINGS OF SEVERAL ALLOYS(1)

The values listed are for rib-and-web forgings with 7-degree draft and standard fillet and corner radii.

Alloy	Minimum Rib Thickness, in., for Forgings of Given Plan Area, sq in.			Minimum Web Thickness, in., for Forgings of Given Plan Area, sq in.		
	Up to 10	10-100	100-300	Up to 10	10-100	100-300
2014 (aluminum)	0.12	0.19		0.10	0.25	0.31
AISI 4340 (low-alloy steel)	0.12	0.19		0.19	0.31	0.50
H-11 (hot-work die steel)	0.12	0.19			0.38	
17-7PH (stainless steel)		0.25			0.38	
A-286 (superalloy)		0.25			0.38	
Ti-6Al-4V (titanium alloy)		0.25			0.31	
Unalloyed Mo		0.38			0.38	

Note: This table is based on data provided by several forging companies. In many cases, the companies did not agree on minimum values for rib widths and web thicknesses. The values presented here indicate the most advanced state of commercial forging practice. There was general agreement on the need for more liberal dimensions with the alloys requiring the greater forging pressures.

When both the metal being forged and the dies are heated to their respective temperatures, thermal expansion causes a change in their dimensions. After forging, the part contracts predictably on cooling as a function of the thermal-expansion coefficient and the final forging temperature. Because of unequal mass distributions, however, the final forging temperature of most forgings is not uniform and shrinkage is variable. This causes length and width variations which must be taken into account in designing the part.

Die Wear. Die wear varies according to the material being forged. For example, the die wear in forging superalloys is usually about twice that for carbon steel. Die-wear tolerances are therefore additive to length and width tolerances.

Die wear is usually greatest at locations in impression dies where there is considerable metal movement, such as at projections and fillets. The die material "wears away" in these areas due to the extreme temperature and pressure conditions. Die wear is greatest when forging metals that form tightly adhering abrasive scale at forging temperatures. Die wear is also accelerated by forging metals at high temperatures because the dies frequently soften from the heat transferred from the workpiece.

Aluminum and magnesium alloys do not cause much die wear; thus, the normal length and width tolerances usually include allowances for any minor wear.

Die Closure. Die-closure tolerances allow for thickness variations of forgings as affected by the closing of forging dies. They represent the variation in any dimension crossing the parting line of the forging and are applied in a direction perpendicular to the major forging plane.

During the forging operation, the die blocks act as thick plates which deflect elastically under the forging loads. Under these circumstances, the dies do not always close uniformly and the forged part may be slightly thicker toward the center. This behavior is more pronounced when forging long, narrow shapes and when forging metals that require high forging pressures. Therefore, die-closure tolerances include normal piece-to-piece variations and also closure variations along the length of a single forging.

Match. Match tolerances (also called "die-shift" and "mismatch" tolerances) apply to the axial alignment of two opposing impression dies. These tolerances are a measure of the lateral displacement of a point in one die from a corresponding point in the opposite die in any direction parallel to the forging plane. In the case of symmetrical hollow forgings, they are called wall-variation tolerances. Since forging machines are essentially large elastic bodies and are subject to variations consistent with the fitting between sliding guides, there is no perfect alignment between the top and bottom die blocks. In order to improve alignment, integral matching "locks" are machined into the opposing dies. The locks provide lead-in surfaces that align the dies during forging. The use of die locks is particularly important when forging unsymmetrical parts that cause side thrust. For symmetrical parts, die locks usually are not essential.

Surface. Surface tolerances indicate the allowable depth of wrinkles, scale pits, and other surface defects on forgings. Surface tolerances depend to some extent on finish

allowances. They are at a minimum for parts retaining as-forged surfaces and at a maximum when finish allowance is provided for machining the surface.

Straightness. Straightness tolerances allow for slight and gradual deviations of surfaces and center lines from the specified contour, which may occur from post-forging operations such as trimming, cooling from forging, and heat treating. Long, thin forgings are more susceptible to warpage than massive shapes; similarly, thin disks warp more than thick disks of comparable diameter. Some alloys like annealed steels, aluminum, and nickel can be cold straightened, while others like magnesium, titanium, and molybdenum usually require warm straightening. Thus, the factors of both shape and alloy have important influences on straightness tolerances.

Flash Extension. Flash-extension tolerances indicate the amount of flash extending from the body of a forging to the trimmed edge of the flash. Trimming generally refers to mechanical shearing of the flash. However, some metals are not amenable to mechanical trimming because of cracking problems, and special tolerances are applied where the flash must be removed by other methods such as torch cutting, machining, or band sawing. The tabulation below lists the most common flash-removal methods for several types of alloys:

Metal or Alloy	Flash Removal Procedures
Aluminum alloys	Usually cold trimmed
Magnesium alloys	Usually band sawed; sometimes warm trimmed
Copper alloys	Hot or cold trimmed, depending on alloy
Armco iron	Cold trimmed
Carbon, alloy, and stainless steels	Hot trimmed
Titanium alloys	Hot trimmed
Nickel alloys	Hot trimmed
Molybdenum alloys	Hot trimmed
Nickel-base superalloys	Small forgings hot trimmed; large forgings generally machined
Tungsten	Usually machined

REFERENCES

(1) Henning, H. J., Spretnak, J. W., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", ASD Interim Report 61-7-876(I), Contract AF 33(600)-42963, (October, 1961), Battelle Memorial Institute.

(2) Dietrich, R. L., and Ansel, G., "Calculation of Press Forging Pressures and Application to Magnesium Forgings", Trans. ASM (1947) V. 38, p. 709-728.

(3) Proffitt, C. E., "Close Tolerance Steel Forging Development", Interim Technical Report on Contract AF 33(600)-36658, (March 24, 1961), Boeing Airplane Company.

(4) Aluminum Company of America, Designing for Alcoa Forgings, Cleveland, Ohio, (1950).

CHAPTER 4

FORGING-DESIGN PRACTICES

ESTABLISHING THE INITIAL FORGING DESIGN

One of the most important considerations in establishing a die-forging design is the total cost of manufacturing the part. For forgings, this cost can be broken down generally into four basic categories:

(1) Material costs

(2) Die costs

(3) Forging costs

(4) Machining costs.

The consumer of forgings realizes that machining and material costs can be reduced by forging close to the finished-part shape and dimensions and, ideally, would like to have forgings made to final form. The forging company, on the other hand, knows that such an approach usually means increasing die costs and increasing forging costs and, therefore, prefers to forge parts designed with a minimum of forging difficulty. Without knowing each other's processing problems, neither the consumer or the forger can establish a compromise design that represents the lowest practical cost. It is therefore important for both the user and producer of forgings to maintain good communications for establishing the most economical design.

Table 4-1 summarizes the information most often needed when establishing an initial forging design. While some of the items listed are not directly related to forging design, they represent factors influencing the final cost of forgings.

Quantity is probably one of the most important items listed. For small quantities, a generously contoured forging with liberal finish allowances and tolerances may be far more economical than a more precise design. The opposite may be true for large quantities. It is a common practice for forging companies to prepare quotations for both types of designs. This approach permits the consumer to make a choice based on total processing costs.

Forging designs are classified by most forging companies into four general categories:

(1) Blocker-type designs

(2) Commercial designs

(3) Close-tolerance designs

(4) Precision designs.

TABLE 4-1. A SUMMARY OF INFORMATION NEEDED WHEN ESTABLISHING AN INITIAL FORGING DESIGN

(1) Name of part – part number

(2) Alloy by name and/or composition – list alternative alloys

(3) Estimated quantity requirements – present and future

(4) Blueprints of finished part

(5) Blueprints of forging design if proposed by purchasers

(6) Finish – specify if machined or unmachined

(a) Prints for machining (specify surface-finish requirements, e.g., rms)

(b) Desired finish envelope if unmachined

(7) Locations for target machining

(8) Metallurgical specifications, e.g., ASTM, SAE, AMS, Military, etc.

(9) Process specifications, if not included in metallurgical specifications

(a) Packing

(b) Rust preventatives

(c) Marking and identification

(d) Special handling

(10) Quality-control requirements, if not included in metallurgical specifications

(a) Nondestructive testing – e.g., hardness, magnetic particle, ultrasonic, dye penetrant, etc.

(b) Destructive testing – types of tests required and frequency

(c) Appropriate testing specifications

(11) Brief description of part and functional requirements – identify locations and direction of maximum stresses

(12) Desired dimensional tolerances

(13) Other information not always covered by specifications: heat treatment, condition, desired hardness range, tooling points

The first three categories represent designs that are progressively closer to the final-part outline and, accordingly, progressively require an increasing number of forging dies and forging steps. The blocker types are characterized by generous contours, large radii, draft angles of 7 degrees or more, and liberal finish allowances. The commercial designs have more refined details, standard draft (5-7 degrees), smaller radii and finish allowances, and specific dimensional tolerances that can be achieved on most commercial forging equipment. Close-tolerance designs are generally considered as those having shallow draft (1-3 degrees), little or no finish allowance, and dimensional tolerances of less than half those for commercial designs. Close-tolerance forgings are normally forged in conventional equipment but usually require added operations such as coining.

The term "precision" as applied to forging designs is not well defined even in the forging industry. Such terms as "close-finish" forging, "draftless" forging, "net" forging, and "close tolerance" forging are often used synonymously with precision forging. For a better definition of these terms, it is necessary to separate basic design elements (draft, fillets, finish allowances, etc.) from tolerances (allowable variations in dimensions). Following this approach the following definitions are suggested:

(1) Close-Tolerance Design - A forging designed with commercially recommended draft, radii, finish allowances, etc., but with dimensional tolerances of less than one-half the commercial tolerances recommended for an otherwise similar part.

(2) Close-Finish Design - A forging designed with the minimum draft, finish allowances, radii, etc., obtainable in conventional forging equipment. Dimensional tolerances on length, width, match, surface, and straightness are about one-half the commercial tolerances, but die wear, flash extension, and die-closure tolerances are about the same.

(3) Close-Finish, Close-Tolerance Design - This design combines the characteristics of (1) and (2).

(4) Precision Design - These designs are either forged or, in some cases, forged and spot machined to precise dimensions with maximum variations on the order of ±0.010 inch.

Precision forging usually requires the use of additional tooling, special forging techniques, and specialized forging machinery. Most precision forgings from alloys other than those of aluminum and magnesium are forged with precise tolerances only on a few reference locations rather than over the entire forging. This permits accurate location in fixtures, jigs, etc., for subsequent machining. In an increasing number of instances, forging companies are spot machining precise tooling-point locations, stops, etc., on conventionally designed forgings. Such forgings then represent a form of precision design.

DESIGNS FOR DISKS, CONES, AND OTHER CIRCULAR SHAPES

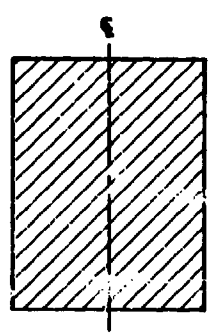
Data sheets are given in this section for each of several forged circular shapes including disks and cones. They contain information on design features (parting-line location, selection of fillet and corner radii, etc.) that have important influences on the

forging operations required. For example, typical designs and forging operations are compared for forging disk shapes from low-alloy steels, titanium alloys, and nickel-base superalloys to demonstrate how both are influenced by the material being forged.

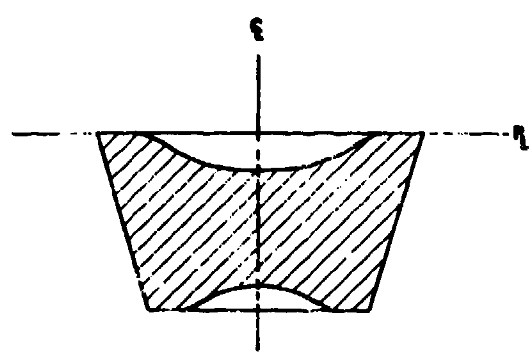
Figure 4-1 shows typical design and forging data for a simple disk shape with a short rim. Such parts are frequently used in jet-engine compressors and turbines. In these applications, the materials may vary from low-alloy steels to titanium alloys to iron-base and nickel-base superalloys, depending on service requirements. These materials represent progressively increasing forging difficulty and require progressively greater material allowances in the webs, ribs, corners, and fillets. In the case of nickel-base superalloys, frequent inspection steps are usually necessary to locate and remove any hot tears that might occur during forging.

Design and forging data for disk shapes with high rims are compared in Figure 4-2 for low-alloy steels and alpha-beta titanium alloys. Two sets of dies (block and finish) are normally required to forge such shapes. In general, the minimum punchout diameter, D_i, is about 3 inches for steel and titanium alloys. Smaller holes may be trimmed, but the center-trim tooling is more complicated (hence, more costly) than that for edge trimming. For this reason, it is often more economical to machine central holes that are smaller than 3-inch diameter unless large production quantities are to be produced. Titanium alloys require more liberal allowances than the steels. Superalloys, especially the second-generation nickel-base alloys such as Udimet 500, Astroloy, René 41, Waspaloy, etc., cannot be forged readily in such designs. Figure 4-3 illustrates how typical superalloy disk and hub forgings appear before and after machining. Particular attention is directed to the generous fillet radii at the rim and central hub regions that are required on the forged shapes.

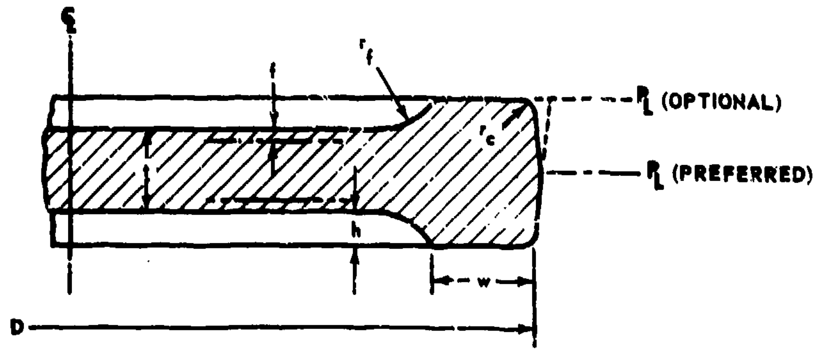
Figure 4-4 shows the typical designs and operations for forging conical shapes from low-alloy steels and unalloyed molybdenum. Because of die-chilling effects and greater forging-load requirements, designs for molybdenum generally require considerably larger section thicknesses than those for low-alloy steels. The recrystallization annealing treatment for molybdenum, denoted as optional, is sometimes used for adjusting final mechanical properties. This also helps to reduce forging-load requirements. Preliminary blocking dies are generally used for molybdenum cones, especially when the $\frac{h}{D}$ ratios exceed 1:1. The blocked forging is generally designed solid with slight depressions on either end, as shown below:



STARTING BILLET



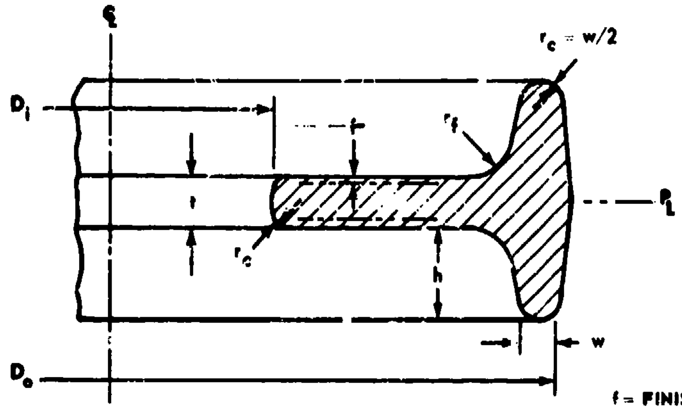
BLOCKED BILLET



f = Finish Allowance

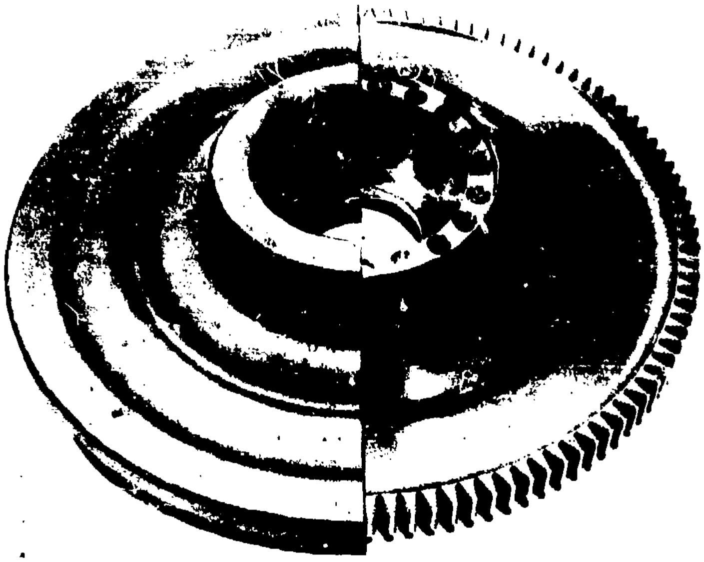
		LOW-ALLOY STEELS D = 20	LOW-ALLOY STEELS D = 10	NICKEL-BASE SUPERALLOYS D = 20	NICKEL-BASE SUPERALLOYS D = 10
TYPICAL MINIMUM DIMENSIONS FOR TWO DISK SIZES	t:	0.50"	0.30"	1.2"	1.0"
	r_f:	0.50"	0.38"	2.0"	1.5"
	r_c:	0.19"	0.12	0.38"	0.25"
	f:	0.12"	0.12"	0.19"	0.19"
	$\frac{h}{w}$ ratio	2:1 max.		1:1 max.	
TYPICAL FORGING OPERATIONS: one set of dies required		Heat Finish forge Trim flash		Heat Upset forge Inspect Heat Finish forge Trim or machine flash	
USUAL FORGING EQUIPMENT:		Drop hammers Hydraulic presses		Drop hammers Hydraulic presses	

FIGURE 4.1. TYPICAL DESIGNS AND PRODUCTION OPERATIONS FOR FORGING SIMPLE DISK SHAPES OF LOW-ALLOY STEELS AND NICKEL-BASE SUPERALLOYS

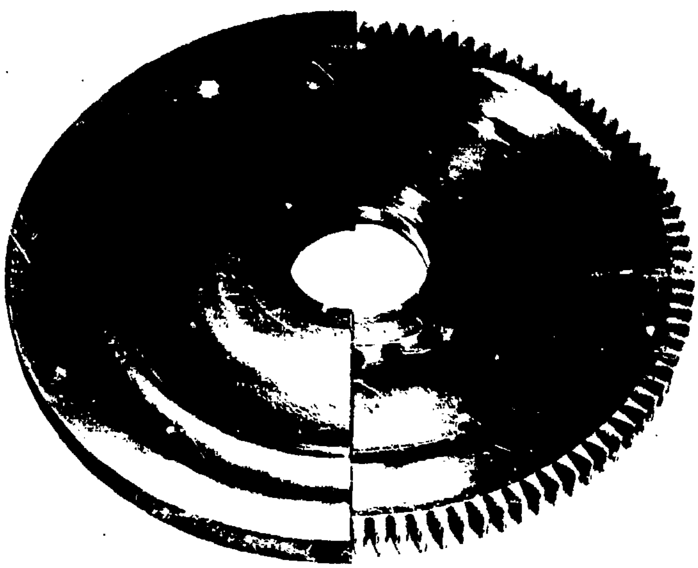


		LOW-ALLOY STEELS		TITANIUM ALLOYS (ALPHA-BETA TYPE)	
		$D_o = 20$	$D_o = 10$	$D_o = 20$	$D_o = 10$
TYPICAL MINIMUM DIMENSIONS FOR TWO DISK SIZES	t:	0.50"	0.38"	0.75"	0.50"
	r_f:	1/6 h	1/6 h	1/3 h	1/3 h
	r_c:	0.12"	0.12"	0.19"	0.12"
	f:	0.12"	0.12"	0.12"	0.09"
	w:	0.50"	0.38"	0.62"	0.50"
	$\frac{h}{w}$ ratio:	4:1 max.		4:1 max.	
TYPICAL FORGING OPERATIONS: (two sets of dies required)		Heat Upset Block Finish forge Trim		Heat Upset Heat Block Inspect Heat Finish forge Trim	
USUAL FORGING EQUIPMENT:		Drop hammers		Drop hammers Hydraulic presses	

FIGURE 4-2. TYPICAL DESIGNS AND PRODUCTION OPERATIONS FOR FORGING HIGH-RIM DISK SHAPES OF LOW-ALLOY STEELS AND TITANIUM-BASE ALLOYS



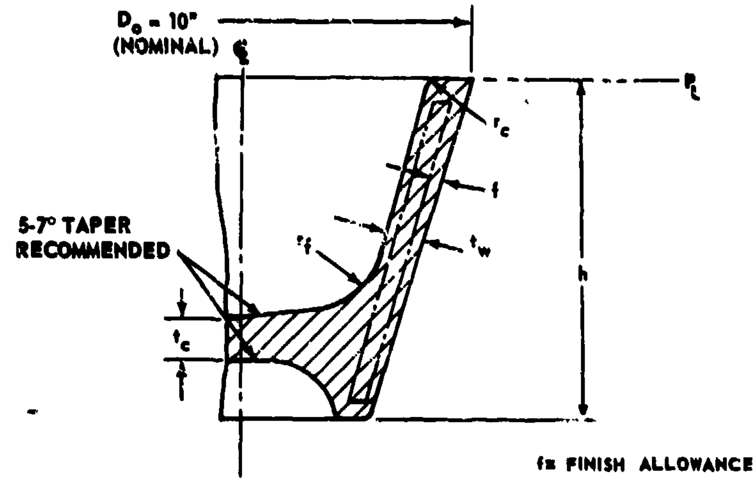
HUB FORGING



DISK FORGING

FIGURE 4-3. TYPICAL APPEARANCE OF SUPERALLOY FORGINGS BEFORE AND AFTER MACHINING

PHOTOGRAPHS COURTESY CAMERON IRON WORKS



		LOW-ALLOY STEEL		MOLYBDENUM	
		$\frac{h}{D_o}$ = 0.5	$\frac{h}{D_o}$ = 1.5	$\frac{h}{D_o}$ = 0.5	$\frac{h}{D_o}$ = 1.5
TYPICAL MINIMUM DIMENSIONS FOR A 10-INCH-OD CONE	t_w:	0.4"	0.5"	0.75"	1.0"
	t_c:	0.5"	1.0"	1.0"	1.2"
	f:	0.12"	0.19"	0.19"	0.25"
	r_f:	1.0"	1.5"	1.0"	1.5"
		(One Set of Dies)		(Two Sets of Dies)	
TYPICAL FORGING OPERATIONS:		Heat Roll corners Upset Finish forge Trim		Heat Upset Block Inspect Anneal (optional) Heat Finish forge Trim	
USUAL FORGING EQUIPMENT:		Drop hammers Hydraulic presses		Drop hammers Hydraulic presses (1-inch minimum thickness)	

FIGURE 4-4. TYPICAL DESIGNS AND PRODUCTION OPERATION FOR FORGING CONICAL SHAPES OF LOW-ALLOY STEELS AND UNALLOYED MOLYBDENUM

When the blocked billet is forged in the finishing die, lateral constraint is provided to the part, thereby reducing the chances for rupturing.

Certain circular shapes require special forging equipment. For example, the venturi-type forging shown in Figure 4-5 requires a forging machine equipped with multiple rams. Such parts are readily forged in multiple-ram presses where the main ram presses the billet between two dies and two opposed side rams simultaneously back extrude the metal to form the venturi shapes (see Chapter 1, Figure 1-9). Since present-day multiple-action machines are slow-moving hydraulic presses, the parts forgeable by this method generally have rather heavy walls and generous finish allowances. It is important to allow sufficient excess material at the ends of these parts because the metal does not always flow uniformly in both directions.

Contoured rings require the use of ring-rolling machines designed for this purpose (see Chapter 1, Figure 1-13). Rolled rings are producible with contours on both inside and outside surfaces. The typical designs and operations for rings with only inside contours are given in Figure 4-6. The inside shape is developed by a comparatively small contoured idler roll. A typical rolled ring of this type is shown in Figure 4-7. Rolled rings with both inside and outside contours require that contours be machined on both the small idler roll and the larger main-drive roll. For this reason, the tooling costs are higher than for rings with inside contours only. Typical designs and production operations for rolled rings with both inside and outside contours are presented in Figure 4-8. A typical ring of this type is shown in Figure 4-9.

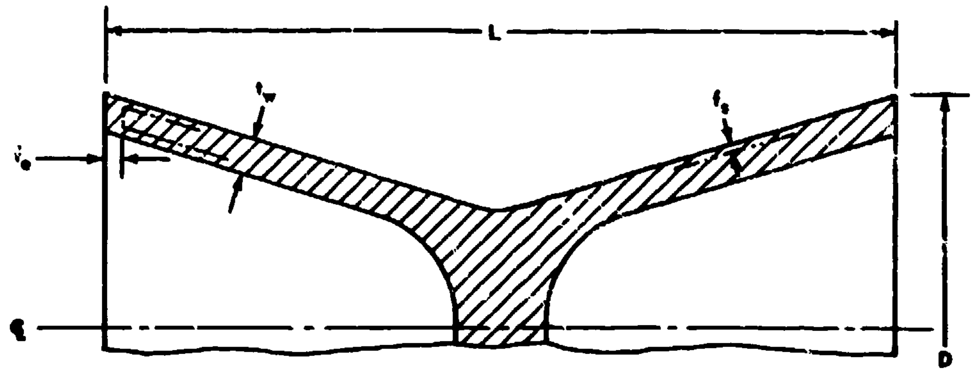
The typical production sequences given in Figures 4-6 and 4-8 for different alloy groups reflect the need for a greater number of operations with materials of increasing forging difficulty. When rolling superalloys, for example, it is quite common to roll in two or more sets of rolls. This is done for two reasons: (1) to minimize cracking by limiting the reductions per roll pass and (2) to provide the controlled amount of reduction necessary for achieving optimum final grain size and mechanical properties.

Figure 4-10 illustrates two large ring shapes that can be successfully contour rolled from alloy and stainless steels. The dimensions given are typical for these alloys. The ring having a moderate contour (Figure 4-10a) is rolled in one set of rolls. The ring of extreme contour (Figure 4-10b), having large differences in diameter top to bottom, requires several roll passes in at least two sets of rolls. Rings of moderate contour may be rolled from certain superalloys and refractory metals, providing the wall thicknesses are greater. Rings producible with the more extreme contours are considered limited to alloy and martensitic stainless steels.

DESIGNS FOR STRUCTURAL SHAPES

Structural shapes are generally thought of as long, narrow parts having ribs, webs, and suitably located bosses for attachment to adjoining structures. The design of forgings for structural shapes is generally more complex than for circular shapes because of the "U" and "H" shaped rib-and-web portions common to most of them.

Data sheets are given in Figures 4-11 through 4-13 to illustrate some of the typical designs and the forging sequences necessary for structural shapes with Type 2, Type 3, and Type 4 ribs (see Figure 3-12). The typical dimensions given are the "average values" reported for commercial designs in an industry-wide survey. Typical designs



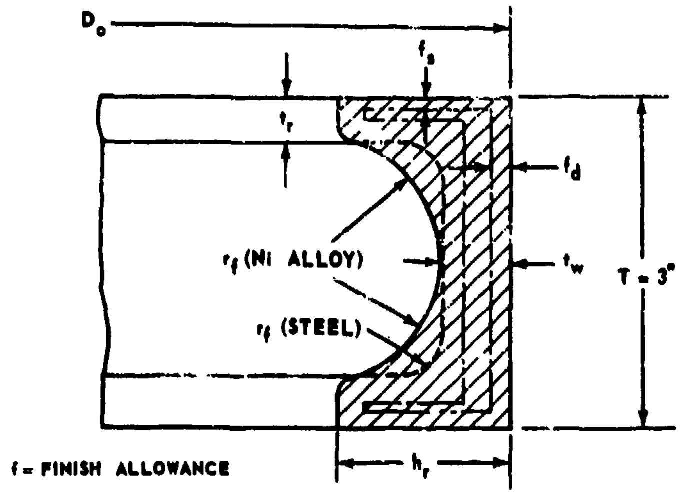
TYPICAL MINIMUM DIMENSIONS FOR THREE SIZES OF VENTURI-TYPE SHAPES		D = 10	D = 15	D = 20
	t_w:	0.75"	1.00"	1.20"
	f_s:	0.25"	0.25"	0.38"
	f_e:	0.38"	0.50"	0.75"

TYPICAL FORGING OPERATIONS:
(one set of dies required)

Heat
Draw stock
Finish forge

USUAL FORGING EQUIPMENT: Multiple-ram hydraulic presses

FIGURE 4-5. TYPICAL DESIGNS AND PRODUCTION OPERATIONS FOR FORGING VENTURI SHAPES OF LOW-ALLOY STEELS



TYPICAL MINIMUM DIMENSIONS FOR VARIOUS RING DIAMETERS

	LOW-ALLOY STEELS		NICKEL-BASE SUPERALLOYS	
	D_o = 10-20	D_o = 20-30	D_o = 10-20	D_o = 20-30
t_w:	0.38"	0.50"	0.50"	1.00"
t_r:	0.38"	0.50"	0.50"	0.75"
f_d:	0.19"	0.25"	0.19"	0.25"
f_s:	0.19"	0.19"	0.19"	0.25"
r_f:	0.25"	0.38"	0.50	1.00"

(Generally, maximum $h_r = T/2$)

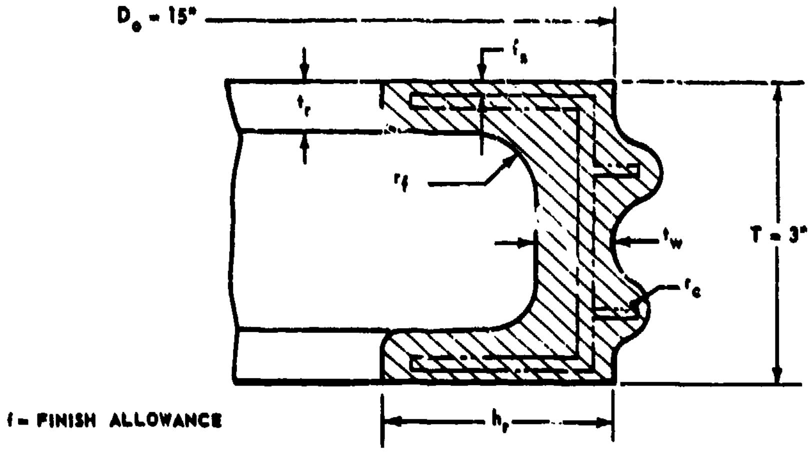
	LOW-ALLOY STEELS	NICKEL-BASE SUPERALLOYS
TYPICAL FORGING OPERATIONS:	Heat Die forge small ring Trim or machine center Heat Finish roll	Heat Upset Inspect Die forge small ring Machine center Inspect Heat Finish roll
USUAL FORGING EQUIPMENT:	Drop hammers Ring rolling machines	Drop hammers Hydraulic presses Ring rolling machines

FIGURE 4-6. TYPICAL DESIGNS AND PRODUCTION OPERATIONS FOR ROLL FORGING RINGS WITH ID CONTOURS OF LOW-ALLOY STEELS AND NICKEL-BASE SUPERALLOYS

Ring shown has been partially rough machined.
Material: Ti-5Al-2.5Sn Alloy

FIGURE 4-7. TYPICAL ROLLED-RING FORGING WITH ID CONTOUR

COURTESY OF LADISH COMPANY



		LOW-ALLOY STEELS	TITANIUM ALLOYS	NICKEL-BASE SUPERALLOYS
TYPICAL MINIMUM DIMENSIONS FOR VARIOUS MATERIALS		D_e = 15	D_e = 15	D_e = 15
	t_w:	0.38"	0.50"	0.50"
	t_r:	0.38"	0.50"	0.50"
	f_s:	0.19"	0.19"	0.19"
	r_f:	0.25"	0.38"	0.50"
	r_c:	0.19"	0.25"	0.38"

(Generally, maximum h_r = T/2)

TYPICAL FORGING OPERATIONS: (starting with forged ring blank)	LOW-ALLOY STEELS	TITANIUM ALLOYS	NICKEL-BASE SUPERALLOYS
	Heat	Heat	Heat
	Finish roll	Rough roll	Rough roll
		Heat	Inspect
		Finish roll	Heat
			Rough roll
			Inspect
			Heat
			Finish roll

FIGURE 4-8. TYPICAL DESIGNS AND PRODUCTION OPERATIONS FOR ROLL FORGING RINGS WITH ID AND OD CONTOURS OF SEVERAL MATERIALS

Ring shown has been completely rough machined
Material: Ti-5Al-2.5Sn Alloy

FIGURE 4-9. TYPICAL ROLLED RING WITH BOTH ID AND OD CONTOURS
COURTESY OF LADISH COMPANY

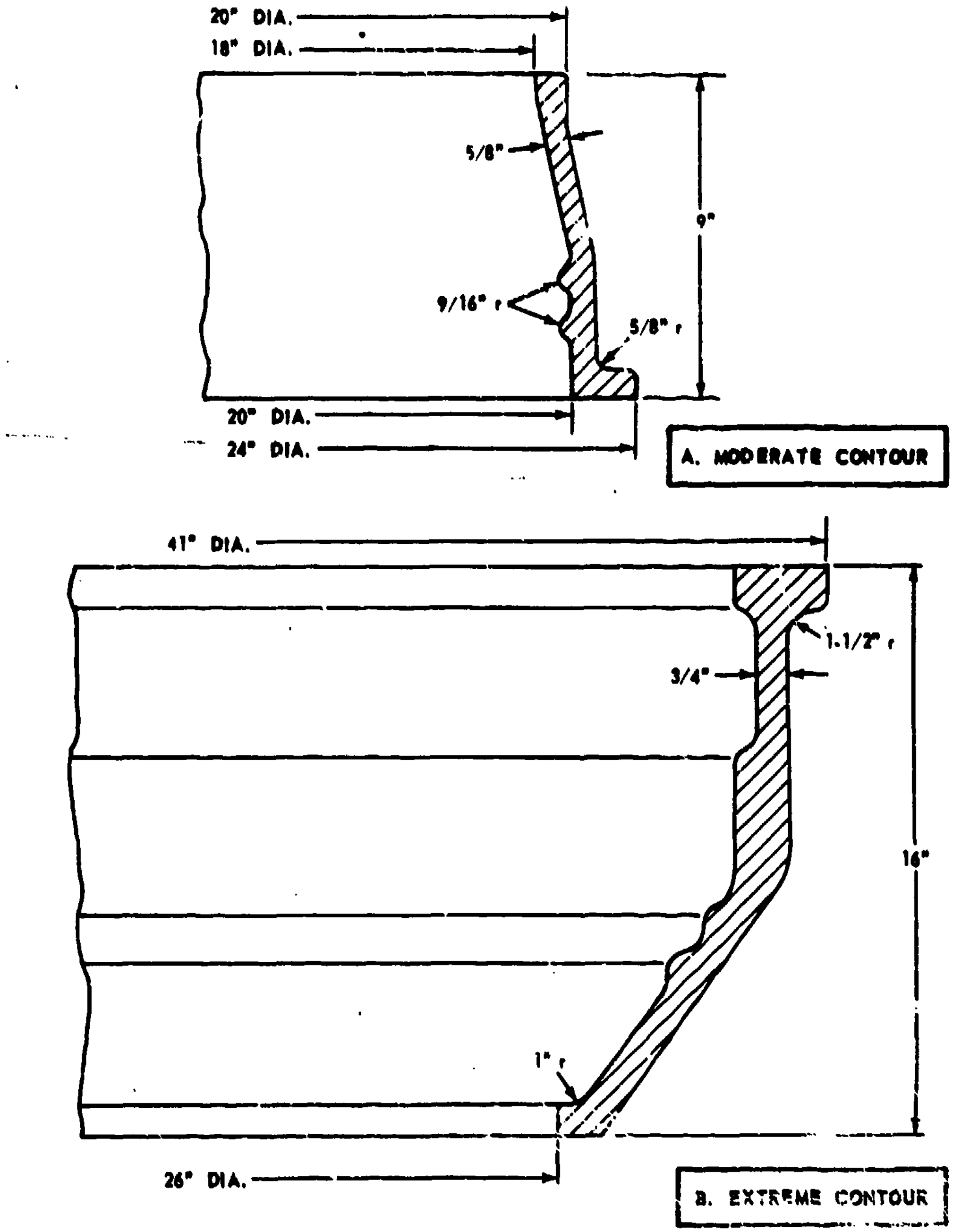
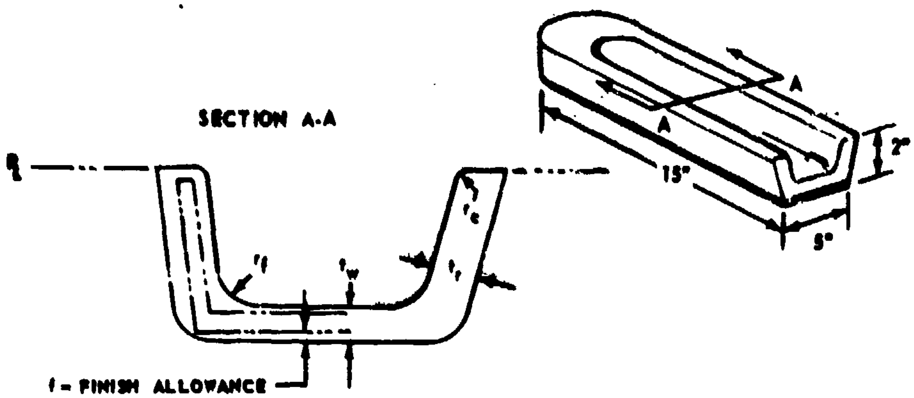


FIGURE 4-10. TYPICAL DIMENSIONS OF LARGE CONTOUR-ROLLED RINGS
OF ALLOY AND STAINLESS STEELS

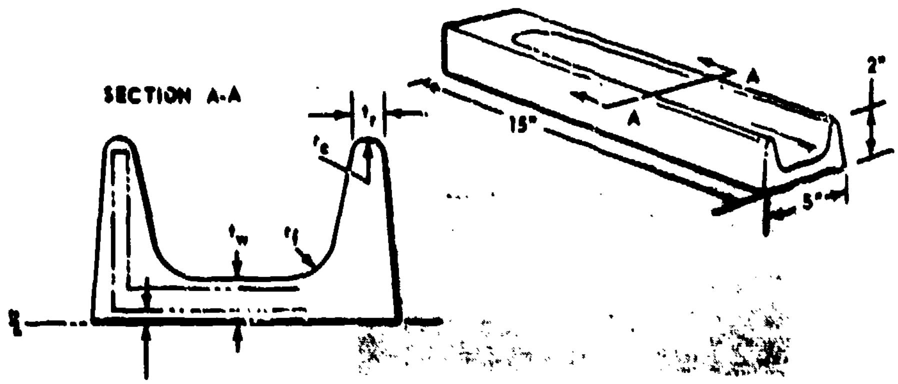


TYPE 2 – PARTING LINE AT TOP OF RIBS

		LOW-ALLOY STEELS	TITANIUM ALLOYS	UNALLOYED MOLYBDENUM	NICKEL-BASE SUPERALLOYS
TYPICAL DIMENSIONS FOR COMMERCIAL DESIGNS	t_r:	0.32"	0.38"	0.50"	0.62"
	t_w:	0.32"	0.38"	0.50"	0.75"
	r_c:	0.12"	0.19"	0.19"	0.25"
	r_f:	0.25"	0.50"	0.50"	0.75"
	f:	0.06"	0.09"	0.12"	0.12"
	Draft:	5°	5°	5°	7°
TYPICAL OPERATIONS:		Heat	Heat	Heat	Heat
		Draw end	Draw end	Draw end	Upset end*
		Flatten	Heat	Heat	Heat
		Finish forge	Flatten*	Flatten	Flatten*
		Trim	Heat	Heat	Heat
			Block*	Block*	Block No. 1*
			Heat	Anneal	Heat
			Finish forge	Heat	Block No. 2*
			Trim	Finish forge	Heat
				Trim	Finish forge
					Machine flash

*In-process inspection usually required.

FIGURE 4-11. TYPICAL DESIGNS AND PRODUCTION OPERATIONS FOR STRUCTURAL FORGINGS OF VARIOUS ALLOYS WITH TYPE 2 RIBS

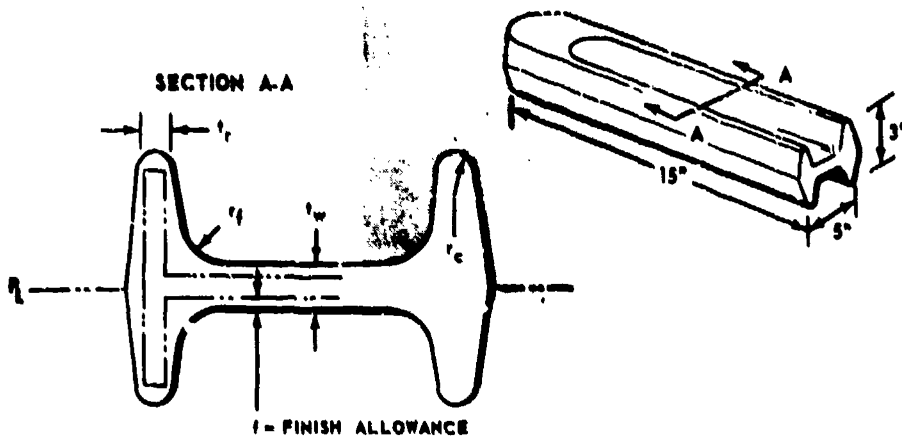


TYPE 3 – WEB AND PARTING LINE AT BASE OF RIBS

		ALUMINUM ALLOYS	LOW-ALLOY STEELS	TITANIUM ALLOYS	IRON-BASE SUPERALLOYS
TYPICAL DIMENSIONS FOR COMMERCIAL DESIGNS	t_r:	0.25"	0.32"	0.38"	0.38"
	t_w:	0.25"	0.32"	0.50"	0.62"
	r_c:	0.09"	0.12"	0.19"	0.25"
	r_f:	0.38"	0.50"	0.75"	1.00"
	f:	0.06"	0.09"	0.12"	0.12"
	Draft:	3°	5°	7°	5°
TYPICAL OPERATIONS:		Heat Draw and Flatten Heat Block Heat Finish forge Trim	Heat Draw and Heat Flatten Block Finish forge Trim	Heat Draw and Heat Flatten* Heat Block* Heat Finish forge Trim	Heat Draw and Heat Flatten Heat Block* Heat Finish forge Trim

*In-process inspection usually required.

FIGURE 4.12. TYPICAL DESIGNS AND PRODUCTION OPERATIONS FOR STRUCTURAL FORGINGS OF VARIOUS ALLOYS WITH TYPE 3 RIBS



TYPE 4 – PARTING LINE AND WEB AT CENTER OF RIBS

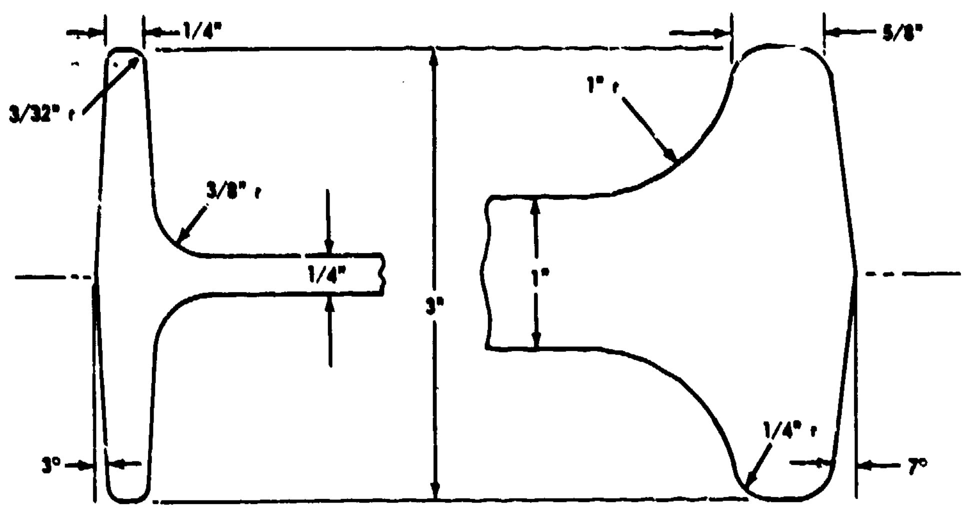
		ALUMINUM ALLOYS	LOW-ALLOY STEELS	IRON-BASE SUPERALLOYS	NICKEL-BASE SUPERALLOYS
TYPICAL DIMENSIONS FOR COMMERCIAL DESIGNS	t_r:	0.25"	0.32"	0.38"	0.62"
	t_w:	0.25"	0.28"	0.62"	1.00"
	r_c:	0.09"	0.12"	0.19"	0.25"
	r_f:	0.38"	0.38"	0.70"	1.00"
	f:	0.04"	0.06"	0.09"	0.12"
	Draft:	3°	5°	7°	7°
TYPICAL OPERATIONS:		Heat Draw one end Flatten Heat Block No. 1* Heat Block No. 2* Heat Finish forge Trim	Heat Draw one end Flatten Heat Block Finish forge Trim	Heat Draw one end Heat Flatten Heat Block* Heat Finish forge Trim	Heat Upset one end Heat Flatten* Heat Block* Anneal (optional) Heat Finish forge Machine flash

*In-process inspection usually required.

FIGURE 4-13. TYPICAL DESIGNS AND PRODUCTION OPERATIONS FOR STRUCTURAL FORGINGS OF VARIOUS ALLOYS WITH TYPE 4 RIBS

are compared for alloys representing increasing levels of forging difficulty. The alloys requiring the most generous contours and the least detail are the nickel-base superalloys. The typical operations listed in the data sheets are not given in precise detail, but rather are presented to illustrate that more operations are required to complete the forging of the more difficult-to-work alloys. Some of the alloys, particularly nickel-base superalloys, often require numerous in-process inspection steps to locate and remove defects.

Figure 4-14 compares the typical design limits for Type 4 rib-and-web cross sections for aluminum alloys and nickel-base superalloys. These shapes represent what may be considered the normal production capability for those two alloy groups. Contours of better shape definition require higher tooling and process-development costs as well as added operations.



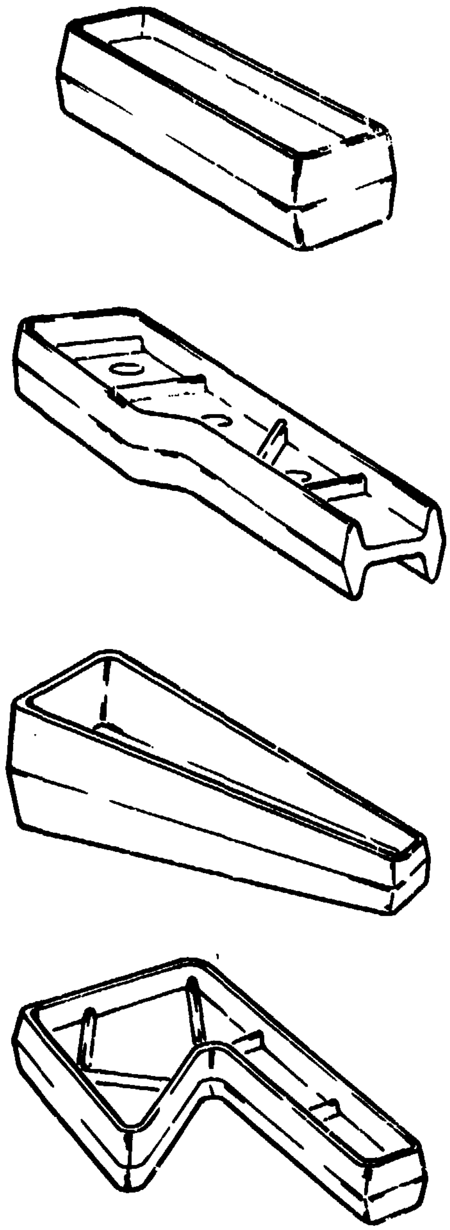
ALUMINUM ALLOYS NICKEL-BASE SUPERALLOYS

FIGURE 4-14. COMPARISON OF TYPICAL DESIGN LIMITS FOR TYPE 4 RIB-AND-WEB STRUCTURAL FORGINGS OF ALUMINUM ALLOYS AND NICKEL-BASE SUPERALLOYS

Of the three types of rib-and-web forging designs illustrated, the Type 2 rib design represents the easiest-to-forge shape. The most common, yet more difficult-to-forge, shape is the Type 4 rib design.

Structural shapes containing cross ribs (Type 1) represent an added level of forging difficulty. A few typical structural shapes are illustrated in Figure 4-15. Such shapes are not produced routinely, and close coordination with forging vendors is generally required to establish suitable designs for these shapes. One of the greatest difficulties with such designs is that of predicting where the metal will flow in the dies

Whenever possible, it is desirable to design these forgings with punch-outs in the webs. This provides a reservoir for any excess metal that otherwise would flow from the web out the flash line and cause defects in the fillet regions similar to the flow-through defects described in Chapter 3.



A. WEB SURROUNDED BY RIBS

B. SERIES OF CROSS RIBS AND WEBS SURROUNDED BY RIBS

C. WEB SURROUNDED BY TAPERED RIBS

D. COMPLEX SHAPE WITH CROSS RIBS AND WEBS OF VARYING SIZES SURROUNDED BY RIBS

FIGURE 4-15. SEVERAL TYPES OF STRUCTURAL FORGINGS THAT REPRESENT INCREASING LEVELS OF FORGING DIFFICULTY

DIFFICULTY INCREASES FROM A TO D

Fillet and corner radii for these designs are about the same as for the Type 4 design. Where ribs intersect, however, the fillet radius at the junction will vary with the angle of intersection. For perpendicular ribs, the fillet radius is generally increased to about 1.5 times the normal radius (the same effect can be achieved by increasing the draft angle in the junction to 10 degrees or more). For angles of intersection progressively smaller than 90 degrees, the fillet radius should be increased further, for example, twice the normal radius for angle of 45 degrees or less. Conversely, for angles of intersection progressively greater than 90 degrees, the fillet radius at the junction approaches the normal fillet radius. Whenever possible, inside cross ribs should be no higher than about one-half the height of the outside ribs. These general rules for designing intersecting ribs on complex structural shapes are illustrated in Figure 4-16. It should be noted that these rules generally apply only to aluminum alloys, magnesium alloys, alloy and stainless steels, and titanium alloys. Because there has been so little need for such forged shapes in refractory metals or superalloys, it can only be assumed that these same design rules would apply for these materials, should the need arise.

DESIGNS FOR FORGINGS WITH PRECISION TOLERANCES

There are two basic methods for producing forgings with precision tolerances. The first consists of forging all details as close as possible to the desired tolerances on dies, followed by final machining. The second method consists of forging to conventional tolerances, then machining specified regions on the forging to close tolerances. This is called "target" machining. The primary aim of both methods is to provide for a minimum amount of machining.

Precision Forging

The concept of precision forging had its beginning with the development of extrusion-type, no-draft aluminum forgings. Forging techniques were developed to the point where tolerances on aluminum rib-and-web shapes could be held to as little as ± 0.005 inch. An example of such a forging is shown in Figure 4-17. Precision forging of aluminum alloys is practical because of several factors:

(1) Forging and die temperatures are essentially the same.

(2) Aluminum does not oxidize significantly.

(3) Forging pressure requirements are low.

(4) Thermal shrinkage is predictable.

(5) Aluminum alloys have excellent forgeability.

Except for magnesium alloys, no other structural material possesses all these advantages.

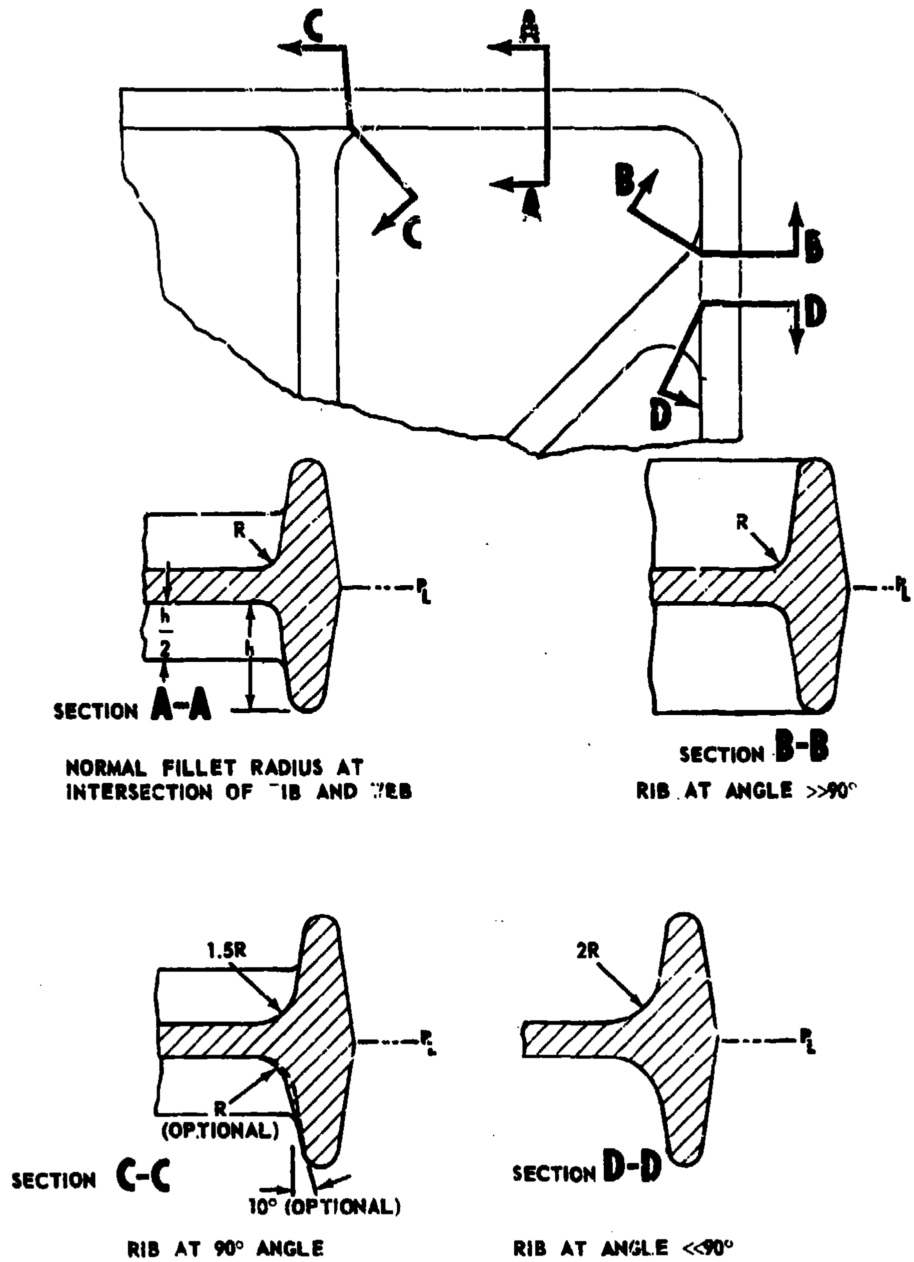


FIGURE 4-16. GENERAL RULES FOR DESIGN OF FILLETS ON STRUCTURAL FORGINGS WITH INTERSECTING RIBS

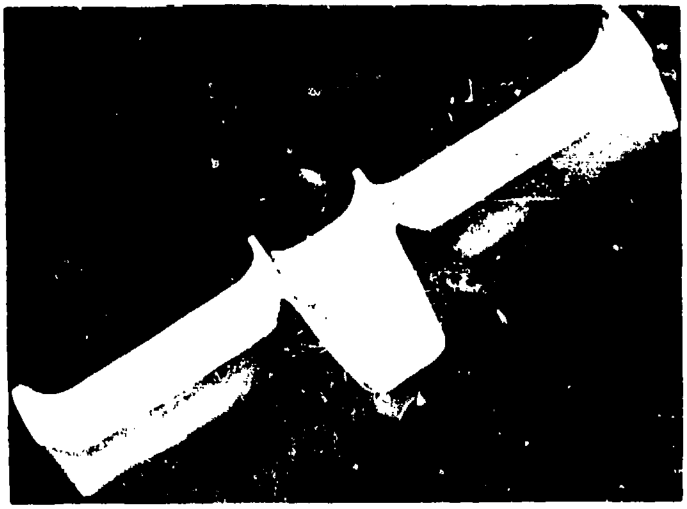


FIGURE 4-17. TYPICAL ALUMINUM-ALLOY PRECISION FORGING

COURTESY OF WYMAN-GORDON COMPANY

It is important to recognize that inherent metal properties determine the capability of forging to small, precise dimensions. Thus, while aluminum and magnesium alloys can be readily forged to extremely close tolerances, it is much more difficult to precision forge steels, titanium alloys, and other materials that require high forging temperatures and high forging pressures, and are inherently less forgeable. Likewise, the more complicated a shape becomes, the more difficult it is to forge to small, precise dimensions.

In the past few years considerable progress has been made toward developing methods for producing precision forgings from metals other than aluminum. Efforts by industry and Government have proven precision-forging capabilities for magnesium alloys, titanium alloys, low-alloy steels, and even tungsten alloys.(1,2,3,4,5,6) The parts successfully forged represent a variety of forging methods and forging-shape complexities. Most of the programs, however, have demonstrated that precision forgings are far more expensive than conventionally designed parts. This is the major reason why there has been comparatively little demand for precision forgings for commercial applications.

Hot-Die Precision Forging. The hot-die forging technique consists essentially of forging in dies heated to the vicinity of the forging temperature. Such forging is normally done in hydraulic presses with slow, controllable strokes. Die temperatures must be maintained within close limits to insure proper metal flow and to maintain dimensional tolerances. The presses are frequently equipped with concentric rams or with special hold-down equipment. The tools are normally equipped with mechanical or pneumatic knockout devices.

In considering the possibility of making a part by hot-die precision-forging techniques, it is important to know certain characteristics of the process:

(1) Die temperatures are limited to about 1000 F for most commercial die materials. Recent experimental work, however, has demonstrated the possibility of die temperatures up to 1600 F with precision-cast dies of superalloys such as Inconel 713 C.(7) A current program exploring the use of refractory metals as die materials may provide a method for using the hot-die technique to temperatures higher than 1600 F.(8) The costs of such tooling would be a determining factor in its selection for a given forging operation.

(2) Die-heating fixtures in close proximity with the dies require careful handling. Special heating fixtures may have to be custom made for each set of forging dies, particularly when die temperatures are in the incandescent range.

(3) Preliminary blanks require dimensional control almost as precise as for the final forging.

(4) Lubricants are generally applied to the forging blanks because the deep, narrow die recesses are difficult to lubricate evenly.

(5) Knockouts must be spaced adequately to permit easy ejection and to prevent warping, but must not interfere with heating fixtures.

(6) Heated dies must have enough strength to withstand pressures on the order of 5 to 10 times the flow stress of the metal being forged.

(7) Since the tooling costs are generally three to five times more than for conventional forgings, the die material should last for several hundred forgings to amortize the initial cost.

Because of the high costs associated with dies heated much above 1000 F, it is likely that the hot-die precision-forging technique will continue to be limited to aluminum and magnesium alloys for forging intricate shapes on a production basis.

<u>Segmented-Die Precision-Forging Techniques</u>(2,3). The most promising method developed for precision forging is the segmented-die technique. This method is particularly suited to the forging of steel parts having thin vertical ribs adjacent to thin webs. Schematic diagrams of the tooling and the operations involved are shown in Figure 4-18. In addition to allowing easy removal of the forging, the method provides for vertical metal flow without risking the flow-through defects common to single-action forging operations on such parts. Several forged parts made by this technique are shown and described in Figures 4-19 through 4-21.

Some of the important requirements and characteristics of the split-die technique are:

(1) The number of steps involved include preliminary die forging to a shape that is designed conventionally but with extremely close tolerances for controlling weight and metal distribution.

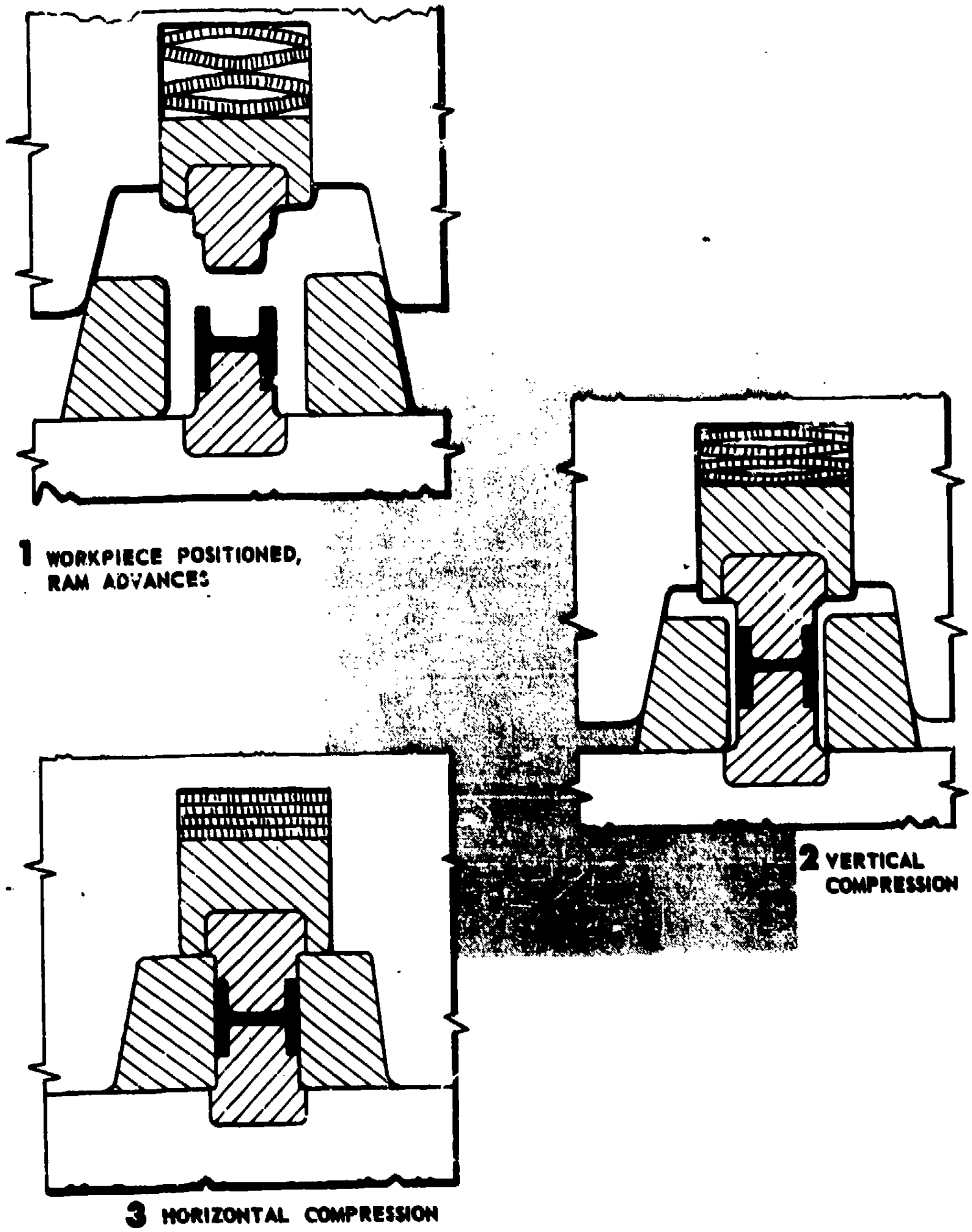


FIGURE 4-18. SCHEMATIC DIAGRAMS OF FORGING SEQUENCE WITH SEGMENTED PRECISION-FORGING DIES

(2) Considerable development is required to establish the proper preliminary range.

(3) The die segments require precise machining, not only on the impression surfaces, but on the mating surfaces.

(4) The reductions obtained by the final sizing operation are quite small and generally localized in nature.

(5) Production rates are comparatively slow.

(6) Dimensional precision is reproducible because the small reductions do not require large amounts of metal movement over die surfaces.

(7) The method is particularly useful for forging steels and titanium alloys.

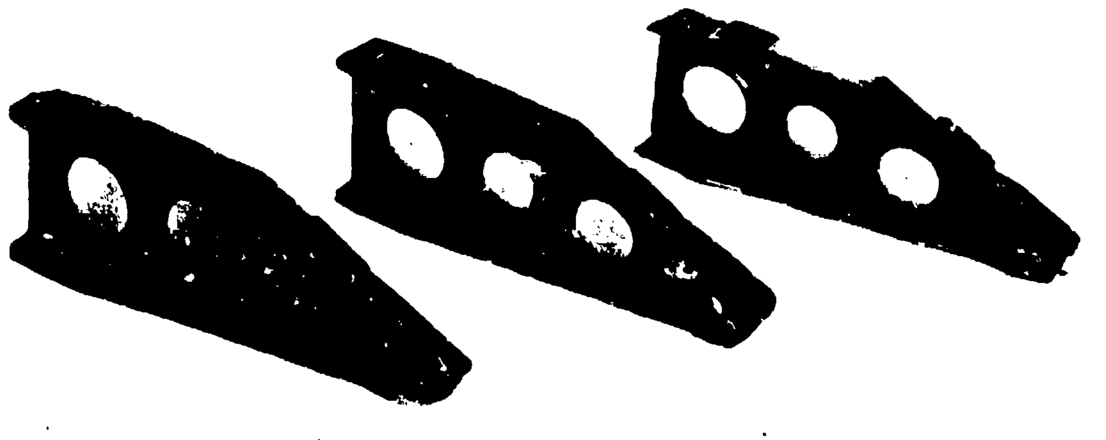
(8) Larger than normal equipment is required to provide the necessary precision.

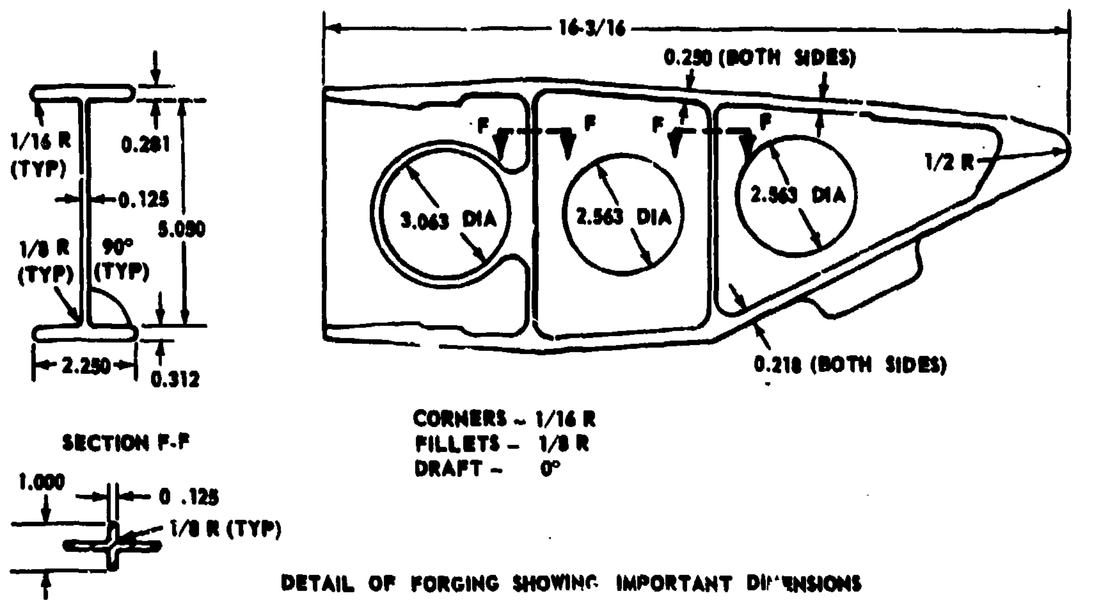
As an example of the last point, the spoiler hinge forging shown in Figure 4-21 required a 5,000-pound hammer for blocking, an 18,000-pound hammer for finishing, and a 4,000-ton hydraulic press for the final sizing operation with segmented dies. Ordinarily, a conventionally designed part could be blocked and finished in a 5,000-pound hammer.

Using rough figures for production rates and equipment costs, the following tabulation illustrates the expected differences in forging costs:

Operation	Equipment Required	Equipment Use Rate, dollars/hr	Forging Rate, pieces/hr	Forging Costs, dollars/part
		Conventional Design		
Block	5000-lb hammer	30	100	$0.30
Finish	5000-lb hammer	30	100	0.30
				$0.60 per forging
		Precision Design		
Block	5000-lb hammer	30	50	$0.60
Finish	18,000-lb hammer	80	20	4.00
Size	4000-ton press	50	20	2.50
				$7.10 per forging

This tabulation does not take into account the higher costs of the precision dies or of other additional operations such as nickel plating, in-process grinding, dimensional inspection, etc., necessary for the precision forging. Even so, the forging costs alone are estimated to be more than ten times those for the conventional design. Thus, the savings in material and machining costs must be quite high to justify the process on an economic basis.





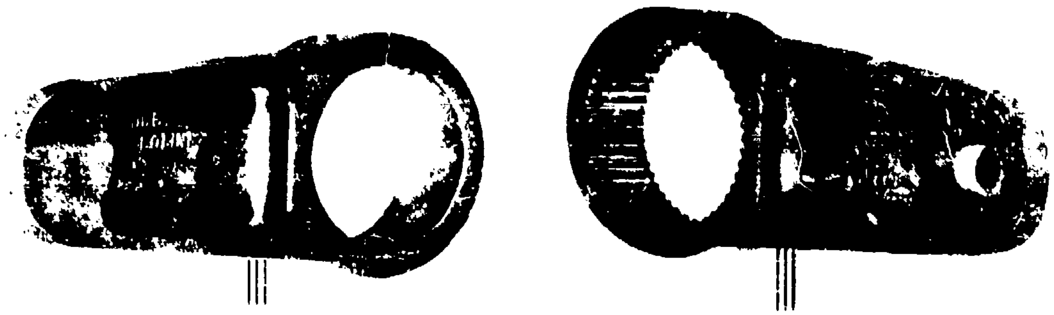
DETAIL OF FORGING SHOWING IMPORTANT DIMENSIONS

Tolerances on Important Dimensions: Rib Height, ±0.010 in.
Web Thickness, ±0.010 in.
Rib Width, ±0.010 in.

Alloys: AISI 4340; Ti-155A; Ti-7Al-4Mo

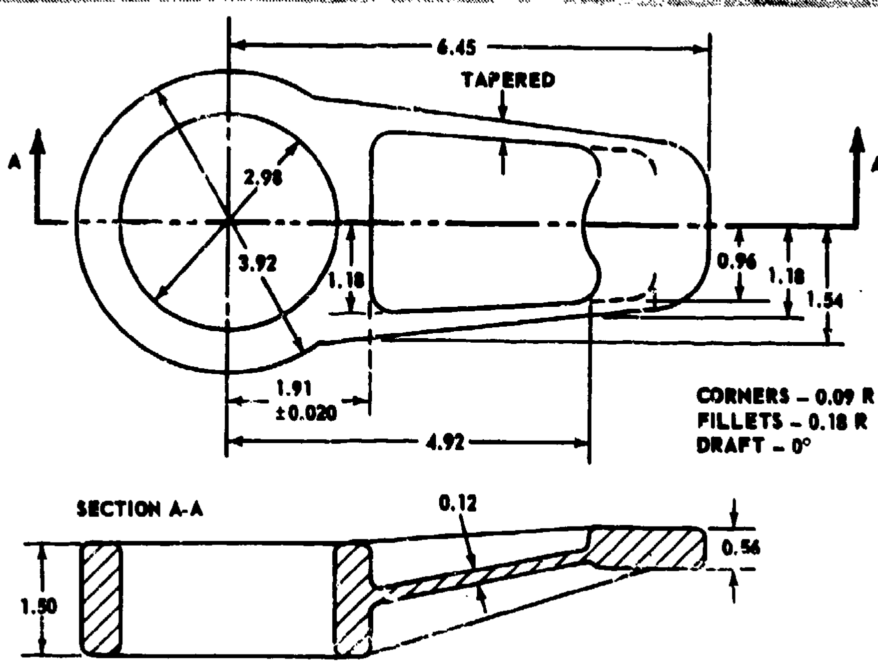
Remarks: Parts forged successfully. Tolerances could not be held on a small run. Metallurgical control difficulties with titanium alloys.

FIGURE 4-19. AILERON HINGE SUPPORT FORGED TO PRECISE DIMENSIONS BY SPLIT-DIE FORGING TECHNIQUE(3)



BELL-CRANK FORGING

MACHINED PART



DETAIL OF FORGING SHOWING IMPORTANT DIMENSIONS

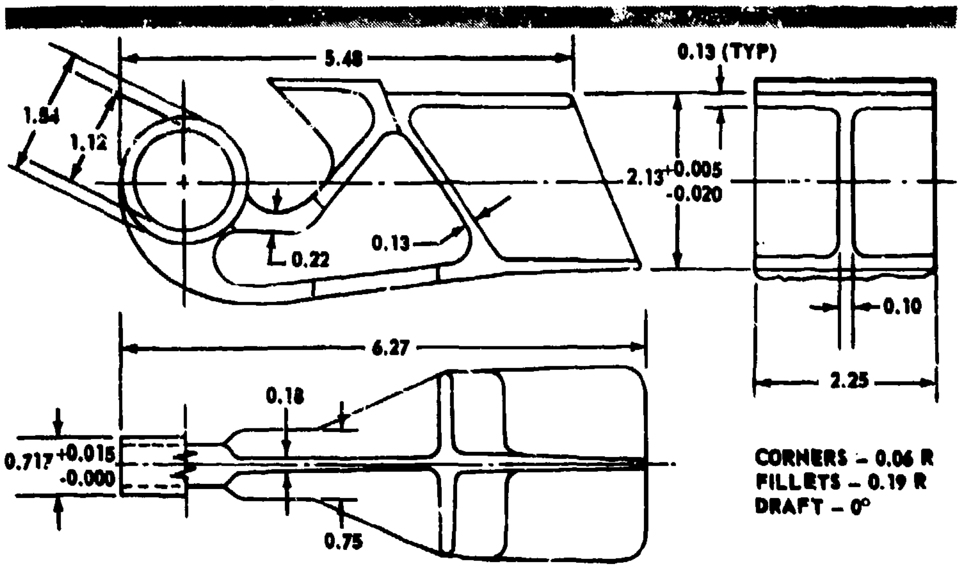
TOLERANCES ON IMPORTANT DIMENSIONS:	Web Thickness,	±0.010"
	Rib Thickness,	±0.010"
	Inside Diameter,	±0.020"

ALLOY: AISI 4340

REMARKS: On a 51-piece run, 81 per cent of dimensions were held within tolerance.

FIGURE 4-20. BELL CRANK FORGED TO PRECISE DIMENSIONS BY SPLIT-DIE FORGING TECHNIQUE[2]

SPOILER-HINGE FORGING



DETAIL OF FORGING SHOWING IMPORTANT DIMENSIONS

TOLERANCES ON IMPORTANT DIMENSIONS:	Web Thickness, 0.10	±0.010"
	0.18	+0.030" −0.010"
	Rib Thickness,	+0.030" −0.010
	Other Dimensions,	±0.030"

ALLOYS: AISI 4340, 300 M, H-11

REMARKS: On a 52-piece run of AISI 4340, 93 per cent of dimensions were held within specified tolerances.

FIGURE 4-21. SPOILER HINGE FORGED TO PRECISE DIMENSIONS BY SPLIT-DIE FORGING TECHNIQUE[2]

High-Energy-Rate Precision Forging.[1,5,6,9] Precision forging by the high-energy-rate technique had its inception in 1958, with the development of the Dynapak high-velocity forging machine which is actuated by compressed gas. Since that time, a fairly wide variety of parts have been made that emphasize the advantages of the process for producing forgings that represent extreme difficulty when forged in slow presses or in multiple-blow hammers. Fine detail and precise dimensions are attainable by high-energy-rate forging because of its basic principle of substituting velocity for mass to develop the energy imparted to a workpiece. The high velocity of forging virtually eliminates die-chilling effects on the workpiece and, thus, extremely thin sections can be forged by this method in a single stroke. At high ram velocities, lighter ram weights and shorter strokes can be utilized, compared to a drop hammer, and die alignment can be maintained with far greater precision. Another important factor contributing to the precision of high-energy-rate forging is its ability to accurately and reproducibly "meter" the energy delivered by adjusting the pressure of the compressed gas.

The closest dimensional precision is obtainable on parts where ribs, bosses, or other projections can be forged in one of the two dies. This automatically eliminates variations from die closure and die shift. Any dimensional variations, therefore, are traceable to either die wear, die sinking, or thermal changes. The forged titanium-alloy impeller shown in Figure 4-22 illustrates this point. Here, the fins and contoured surfaces are formed in the bottom die. On the opposite side of the parting line, the part is designed with conventional finish allowances. The threaded shank is forged on the back for removing the part from the lower die when the dies separate. This minimizes the risk of die softening and helps to increase die life. Because the dimensions on the upper die are not critical, die wear and thermal softening are not significant problems there. Figure 4-23 shows another part designed on this same principle. This part was forged in one blow from AISI 4340 steel.

If dimensional precision is required in both dies and across the parting line, tolerances must necessarily be broader to include die-closure and die-shift variations. Also, the die chosen for gripping the workpiece will exhibit greater dimensional variations since it will be subjected to thermal softening and greater wear. It has been reported that the "gripping die" will wear out three to five times faster than the opposite die if similar precision is required in each. The starting blanks must be held to close dimensional tolerances, and process variables such as forging temperatures, lubrication, die temperatures, and transfer time must be controlled within close limits. If knockout arrangements are used for removing forgings from a die, it is important that the part repeatedly stays in that die.

Some of the important characteristics and limitations of high-energy-rate forging that must be taken into account in considering the process for making a given part are:

(1) Significant temperature increases caused by the rapid deformation may result in incipient melting when forging certain alloys (e.g., high-carbon steels, high-strength aluminum alloys, superalloys), and cause adverse phase changes in others (e.g., beta embrittlement in titanium alloys).

(2) Metals exhibiting low ductility under rapid deformation rates cannot be forged readily by high-energy-rate methods. Magnesium and beryllium alloys are noted for this behavior.

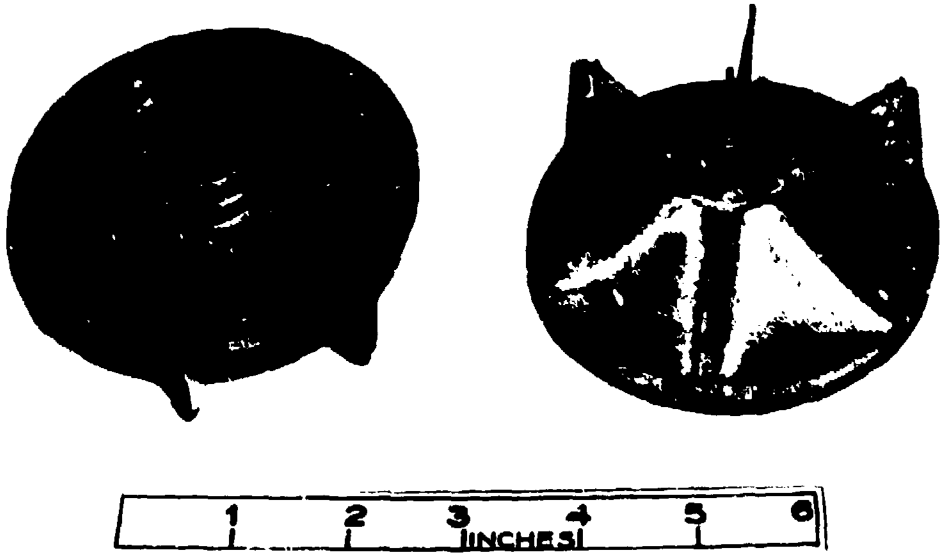


FIGURE 4-22. TITANIUM-ALLOY IMPELLER PRECISION FORGED ON A HIGH-ENERGY-RATE MACHINE (Ti-6Al-4V ALLOY)

COURTESY OF GENERAL DYNAMICS CORPORATION

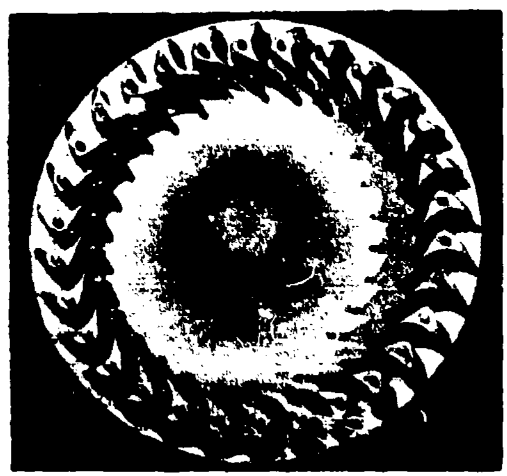


FIGURE 4-23. AISI 4340 TURBINE WHEEL FORGED ON A HIGH-ENERGY-RATE MACHINE

COURTESY OF GENERAL DYNAMICS CORPORATION

(3) The technique is particularly suited to forging alloys that require high forging temperatures and that are not influenced greatly by rapid deformation (e.g., steels, refractory metals, and nickel alloys having broad forging temperature ranges).

(4) Part size, at present, is generally restricted to about 25-30 square inches of plan area.

(5) Part shape is generally restricted to parts that are symmetrical about some axis or centerline. Parts difficult to define geometrically require extreme caution to avoid side thrust during forging.

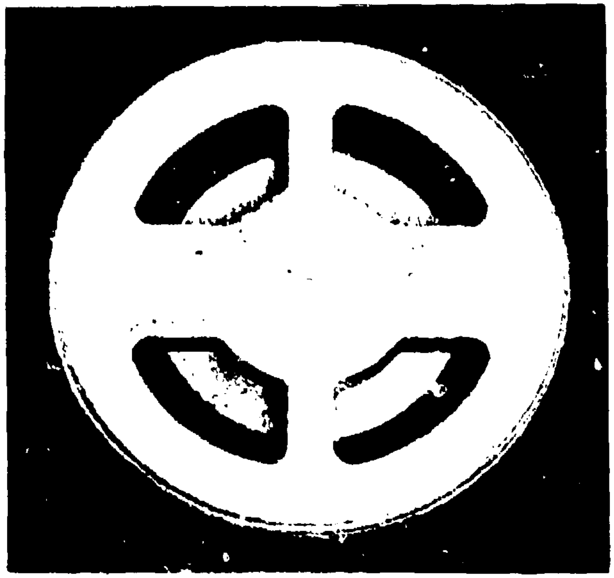
Figures 4-24 through 4-27 show several typical parts that have been forged by high-energy-rate techniques and illustrate the important design features. Although these parts are not all considered precision forgings, they display designs that are extremely difficult or impractical to produce by conventional forging methods.

Target Machining

With the possible exception of the small parts producible by high-energy-rate techniques, the majority of forgings are far more costly when forged to precision tolerances than when forged to commercial designs and tolerances. Yet precise dimensions are necessary because of the requirements of automatic machining. In many instances, these precise dimensions are needed only on a few locations on a forging to permit reproducible fixturing. A useful method for accomplishing this is target machining.

Ordinarily, forgings are inspected by the forging company to be sure the final part can be machined. This inspection is generally repeated at the consumer's facility before the part is released for machining. The machine shop will again lay out the forging to establish the locations for the first machining cut, after which the part is machined. This double and triple inspection can be avoided by having the forging company make the first machining cut in prescribed locations. Only one dimensional inspection is necessary. Upon laying out the forging, the vendor essentially "locates" the final part within the forging envelope. The first machining cut can be a lathe-turned flange on a disk shape, a milled flat surface on a fitting, a drilled centerhole on a cylinder, or a broached slot on a more complex shape. This technique has been used for many years on forged crank shafts but has only recently become popular for other forging shapes.

Substantial cost savings can be realized by eliminating the extra inspection and by permitting the use of less costly forging designs. Thus, an important step for establishing an economical forging design is to review the fixturing requirements and determine whether the forging vendor can make the first machining cut.

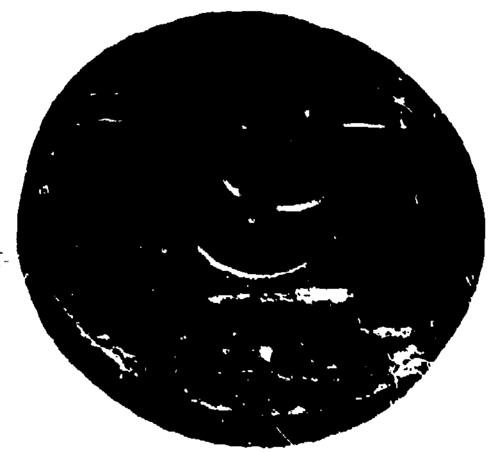


Part: Port Plate (approximately 4 in. in diameter)
Material: Nitralloy G
Operations: Forged in one blow at 2100 F on Model 1210 Dynapak, 525-psi fire pressure, 6-in. stroke
Design: Deep pockets forged with zero draft and 1/32-inch fillet and corner radii. Tolerances on ports, ±0.005 in.

N83012

FIGURE 4-24. PORT PLATE PRECISION FORGED ON A HIGH-ENERGY-RATE MACHINE

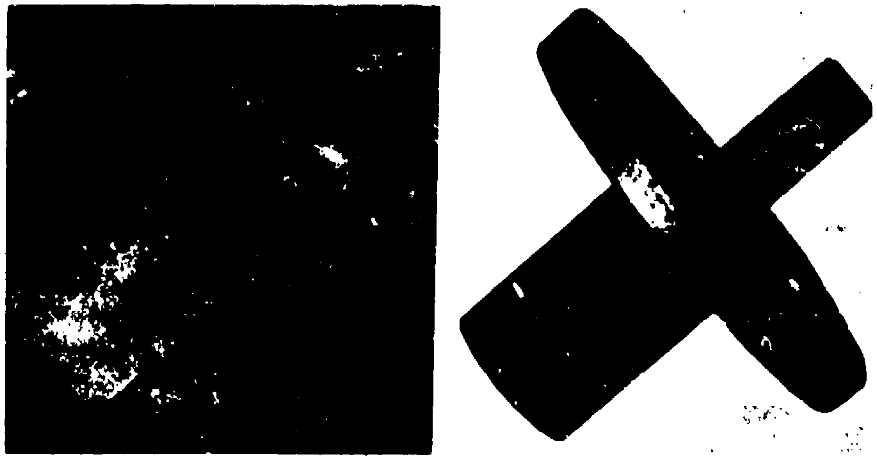
COURTESY OF BENDIX CORPORATION



Part: Spline Plate (approximately 3-1/4 in. in diameter)
Material: AISI 3740 steel
Operations: Forged in one blow at 2150 F on Model 1210 Dynapak, 525-psi fire pressure
Design: Pockets forged with zero draft

FIGURE 4-25. SPLINE PLATE PRECISION FORGED ON A HIGH-ENERGY-RATE MACHINE

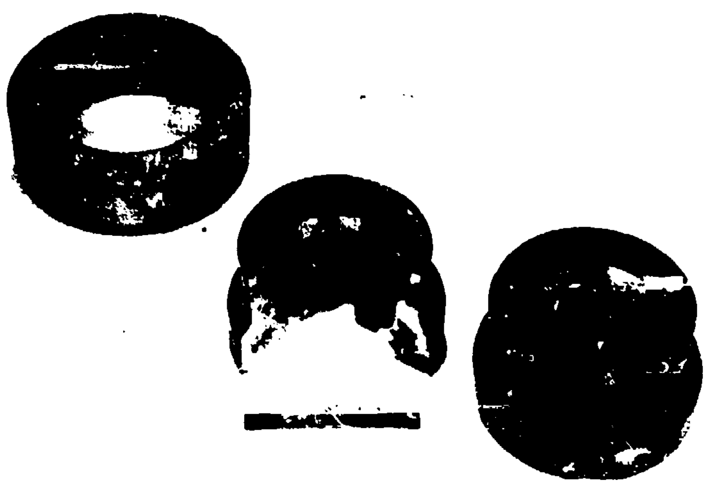
COURTESY OF BENDIX CORPORATION



Part: Flanged Shaft (approximately 3-3/4 in. in diameter)
Material: AISI 9310 steel
Operations: Forged in one blow at 2100 F on Model 1210 Dynapak, 900-psi fire pressure, 7-in. stroke
Design: Shafts forged with zero draft. Web thickness is 1/8 inch, with tolerances of ±0.010 inch.

FIGURE 4-26. FLANGED SHAFT PRECISION FORGED ON A HIGH-ENERGY-RATE MACHINE

COURTESY OF BENDIX CORPORATION



Part: Pipe Union (approximately 3 in. in diameter)
Material: C1015 steel
Operations: Forged in one blow on Model 1210 Dynapak from billet cut from seamless heavy-wall pipe
Design: ID and OD of reduced section forged with zero draft.

FIGURE 4-27. PIPE UNION PRECISION FORGED ON A HIGH-ENERGY-RATE MACHINE

COURTESY OF BENDIX CORPORATION

REFERENCES

(1) Henning, H. J., Spretnak, J. W., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", ASD Interim Report 61-7-876(3) on Contract AF 33(600)-42963, Battelle Memorial Institute (May, 1962).

(2) Proffitt, C. E., "Close Tolerance Steel Forging Development", Boeing Airplane Company, Contract AF 33(600)-36659, Final Technical Engineering Report to ASD (April, 1962).

(3) Hicks, J. S., "Establishment of Production Methods for Aircraft Precision Forgings in Titanium Alloys", North American Aviation, Final Engineering Report on Contract No. AF 33(600)-33703 (November, 1959).

(4) Lake, F. N., and Breznyak, E. J., "Tungsten Forging Development Program", Thompson Ramo Wooldridge, Inc., Progress Reports on Contract AF 33(600)-41629.

(5) Palsulich, J. M., and Headman, M. L., "Developments in High-Energy-Rate Forging", Paper presented at the ASTME Creative Manufacturing Seminar (September, 1961).

(6) Riordan, T. S., "High-Energy-Rate Forming in Production at Bendix-Utica", Paper presented at the ASTME Creative Manufacturing Seminar (September, 1961).

(7) Nickols, H. R., Graft, W. H., Pulsifer, V., and Gouwens, P. R., "Development of High-Temperature Die Materials", Armour Research Foundation, Chicago, Illinois, AMC Technical Report 59-7-579, on Contract AF 33(600)-35914 (June 15, 1959).

(8) Armour Research Foundation "Development of 2400 F Forging Die System", Interim Technical Progress Reports on Contract AF 33(600)-42861.

(9) McDaniel, J. L., "Pneumatic Mechanical Forming", Paper presented at the ASTME Creative Manufacturing Seminar, Detroit, Michigan (1961).

CHAPTER 5

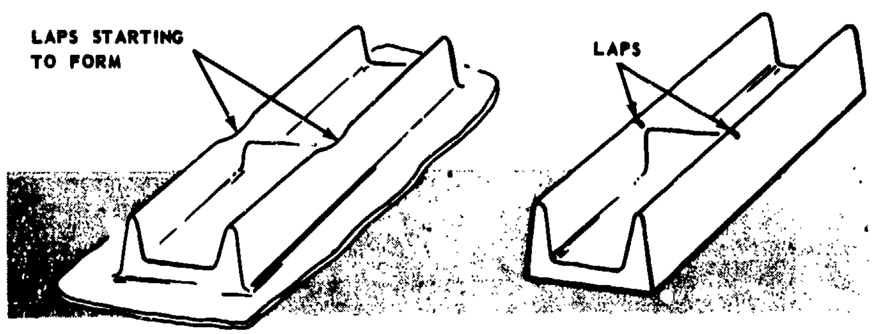
FORGING DEFECTS

CLASSIFICATION OF DEFECTS

Forging defects are defined as those that result from improper forging operations. Certain material defects, such as pipe, seams, and segregation also can result in defects in forging, but this chapter is concerned only with defects associated with the forging operation itself. Material defects are described in later chapters dealing with specific forging alloy systems.

Laps

The term "lap" includes a multitude of defects that form whenever metal folds over itself during die forging. Laps are found most frequently where vertical and horizontal sections intersect. In these cases, the causes are usually traceable to improper selection of fillet radii. Metal flowing nonuniformly in vertical cavities may form a lap when the metal finally fills the cavity. This is a particular problem when the vertical sections of a forging vary significantly in volume requirements. Figure 5-1 shows a forging containing two side ribs and a cross rib in two stages of forging: partially filled and completely filled. A lap often forms in the side rib at the location corresponding to the cross rib where the cross rib draws metal away from the side ribs.



Partially Forged Part　　　　Finished Part

FIGURE 5-1. TYPICAL LAPS FORMED WHEN METAL FLOWS NONUNIFORMLY IN A RIB-AND-WEB FORGING

Laps can also occur during such preliminary forging operations as swaging, rolling, edging, and fullering. Such laps generally show up on die forgings at a specific distance from the end of the die but at a random location about the horizontal centerline.

Coarse-Grain Wrinkles

Forging billets containing coarse grains, whether as-cast or wrought, will develop wrinkles during forging. An example of such wrinkles is shown in Figure 5-2. Shop terms commonly applied to this type of forging defect are "orange peel" or "elephant skin". When such billets are forged in closed dies, these wrinkles often fold in to cause a series of small laps. Although they are seldom very deep, these laps produce a poor surface appearance that often necessitates considerable grinding and restrike forging.

FIGURE 5-2. TYPICAL APPEARANCE OF UPSET FORGED BILLET SHOWING WRINKLES THAT ARE CAUSED BY COARSE GRAINS

Flow-Through Defects

Flow-through defects are essentially laps that form when metal flows past die recesses after they have filled. Figure 5-3 shows, for example, how a completely filled, sound rib-and-web forging can develop flow-through defects by continuing to forge after filling is complete. An important point here is that, although the lap may be shallow, an undesirable grain-flow pattern may extend well below the surface and lead to unsatisfactory mechanical properties in the vertical rib.

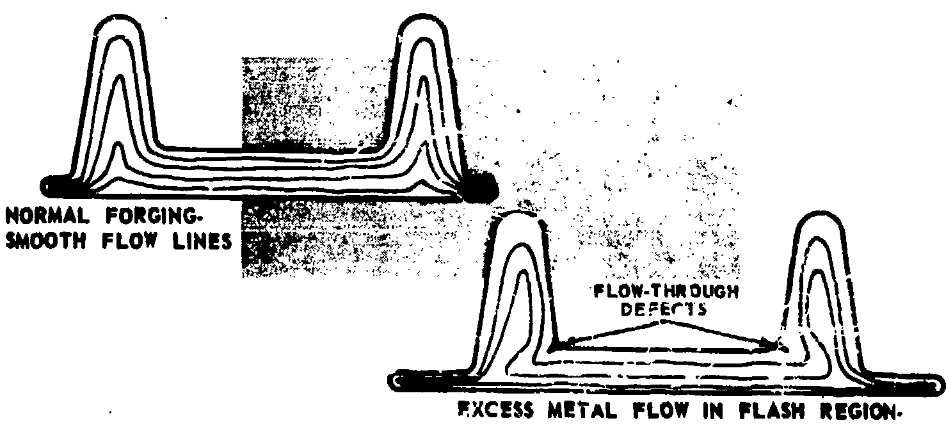


FIGURE 5-3. TYPICAL CAUSE OF FLOW-THROUGH DEFECTS

Flow-through defects can also occur even when the die impression is not completely filled. This happens most often when the metal in the rib or projection exhibits an increasing resistance to flow due to work hardening or die-chilling effects. This is one of the main reasons for using large fillet radii on forging designs for refractory metals or superalloys.

Flow-through defects can also occur when trapped lubricant forces metal to flow past an impression. A frequently used shop term for this is "blowthrough" when the trapped lubricant flows with the metal and is forced into the lap.

Hot Tears and Center Bursts

Hot tears are surface defects that occur when metals rupture during forging. Examples of such defects are shown in Figures 5-4 and 5-5. The presence of segregation, seams, or low-melting or brittle second phases promotes hot tearing at the surface of a forging during upsetting.

Center bursts are ruptures that occur in the center of billets and are most frequently encountered when side forging. Center bursts of the kind shown in Figure 5-6 are usually traceable to segregation and brittle second phases. Center bursts sometimes occur when the temperature of a metal increases significantly as a result of large, rapid reductions and incipient melting occurs. Nickel-base and magnesium-base alloys are particularly sensitive to center bursts from this cause.

Disturbed-Metal Defects

Defects formed when surface metal shears away from the subsurface metal are called "disturbed metal" defects. They are usually caused by either too much or too little lubrication applied during die forging. These defects are a particular problem in forging aluminum and magnesium alloys where there are no loose oxides to provide a parting action between the dies and workpiece. The defects are usually quite shallow but they cause an undesirable surface appearance. Other shop terms applied to disturbed-metal defects are "oil scrub" for those traceable to surplus lubrication, and "dry die" for those traceable to seizing and galling from inadequate lubrication. The dry-die defects are sometimes found on forgings from nickel-base and titanium-base alloys. Again, with these materials, there is no loose oxide that would help minimize metal-to-metal contact.

Thermal Cracks

Cracks caused by stresses resulting from nonuniform temperatures within a metal are called "thermal cracks". In the case of light sections of high-hardenability steels, for example, cracks formed on cooling originate at the surface and extend into the body of the forging. Thermal cracks of this type can be avoided by cooling forgings slowly either in insulating material or in a furnace.

Another type of thermal crack can develop when forgings are heated too rapidly, as illustrated in Figure 5-7. The internal ruptures form because the hotter surface layers expand more than the cooler metal near the center. Since the tensile stress developed at the center depends on the temperature gradient, such cracks are more likely

FIGURE 5-4. TYPICAL APPEARANCE OF HOT TEARS ON BILLETS UPSET WITH FLAT DIES

FIGURE 5-5. TYPICAL APPEARANCE OF HOT TEARS ON CLOSED-DIE FORGINGS

FIGURE 5-6. TYPICAL APPEARANCE OF CENTER BURST OCCURRING

to be encountered in metals with poor thermal diffusivity. Higher coefficients of thermal expansion are also unfavorable. Cracks of this kind occur more frequently in forgings having section sizes larger than about 3 inches. The cracks may be avoided by either of the following measures:

(1) After forging, reheat the part in the forge furnace to remove most of cold-work stresses, then air cool.

(2) If the forgings are cooled to room temperature after forging, reheat slowly through the range of 1000 F to 1500 F before solution heat treating.

The second method is the usual choice for many materials because of possible problems of grain-size control with the first. It should be noted that the stresses causing the cracks may otherwise cause sufficient strain to promote abnormal grain growth on heating. As can be seen in Figure 5-7, the die forging contains evidence of coarse grains, particularly in the central portions. It appears that grain growth occurred in this forging, after the cracks developed, upon continuous heating to the normal solution annealing temperatures.

FIGURE 5-7. TYPICAL THERMAL CRACKS THAT CAN OCCUR IN LARGE FORGINGS WHEN HEATED TOO RAPIDLY

INSPECTION METHODS

The surfaces of die forgings are generally roughened to some extent by either sand blasting, tumbling, pickling, or scaling. For this reason, external defects are not always observable with the naked eye and nondestructive testing procedures are needed for locating surface defects. The most common methods for detecting these defects are:

(1) Magnetic-particle inspection for magnetic materials

(2) Dye-penetrant inspection for nonmagnetic materials

(3) Fluorescent-penetrant inspection for nonmagnetic materials

(4) Eddy-current inspection for both magnetic and nonmagnetic materials.

Once detected, external defects must be identified accurately to minimize future occurrences.

Laps that are caused by improper die design are generally identifiable since they normally occur in the same place on each forging. Although thermal cracks are sometimes mistaken for laps, laps can be distinguished readily by microscopic examinations. They are usually characterized by differing flow patterns on either side and they frequently contain oxides, lubricants, and other contaminants that were enveloped during forging. Opposite sides of laps usually appear smooth from the sliding of one surface over another.

CHAPTER 6

FORGEABILITY OF METALS AND ALLOYS

DEFINITION OF FORGEABILITY

Investigators do not agree on the meaning of the word "forgeability". Some use the the term forgeability to denote a combination of both resistance to deformation and plasticity, while others hold that the term should be restricted to describing the ability of a metal to undergo deformation without fracture. To avoid these different connotations, another group of investigators has used the term "malleability" to describe the latter characteristic. It is common practice in the forging industry to use still different terms to describe metals. It might be said, for example, that "a metal has good forgeability, but it is stiff".

Forgeability is defined here as the tolerance of a metal or alloy for deformation without failure, regardless of forging-pressure requirements.

Most alloys can be classified into one of three groups, i. e., those showing (1) good forgeability, (2) poor forgeability, and (3) variable forgeability. Alloys such as AISI 4340, Ti-6Al-4V, or Type 304 stainless steel always show good forgeability under proper forging conditions. On the other hand, arc-cast alloys of tungsten and molybdenum show poor forgeability unless the coarse-grained ingots are first broken down by extrusion. The third group of alloys comprises those which are likely to demonstrate good forgeability in one heat and poor in the next, because of small differences in composition or microstructure. Some of the high-chromium and chromium-nickel stainless steels show this variable forgeability. In these cases, the variation is usually traceable to excessive amounts of delta ferrite in the microstructure. Essentially single-phase alloys containing significant amounts of grain-boundary phases (e. g., oxides in molybdenum) will usually exhibit poor forgeability, especially if the grain size is large.

Forgeability evaluations at a particular temperature do not necessarily define the ease with which a metal can be forged in impression dies under shop conditions. In die-forging operations, metal temperatures usually vary because of die chilling and because of energy absorption, which causes heating. Consequently, a metal with a wide forging temperature range may be easier to forge than is one which withstands equal amounts of deformation without rupture in standardized tests.

MEASUREMENT OF FORGEABILITY

Forgeability evaluations reported by most investigators are based on one or more of six testing methods. In the general descriptions of these tests, which follow, an attempt has been made to show where each test can be used most effectively.

Hot-Twist Test. This test consists of twisting a metal at elevated temperatures and counting the number of twists to failure; a larger number of twists before failure indicates better forgeability. This test is particularly useful for evaluating alloys that are normally hot worked. The twisting of alloys at cold-working temperatures is complicated

by strain-hardening effects, which cause misleading results that do not reflect the true deformation tolerance of the particular alloy.

A schematic diagram of a hot-twist testing apparatus is shown in Figure 6-1. Several investigators[1,2,3,4,5] have used devices similar to this. Clark and Russ[4] modified the apparatus to measure both revolutions to failure and torque as comparative values for forgeability and flow stress, respectively, at various temperatures. Data obtained on both properties are useful for estimating optimum forging temperatures. A report by Tout and Banning[5] provides useful details on how variables in the twist test influence the results obtained. In general, both the number of twists to failure and the torque required for twisting increase with increasing rotational speeds at any given temperature.

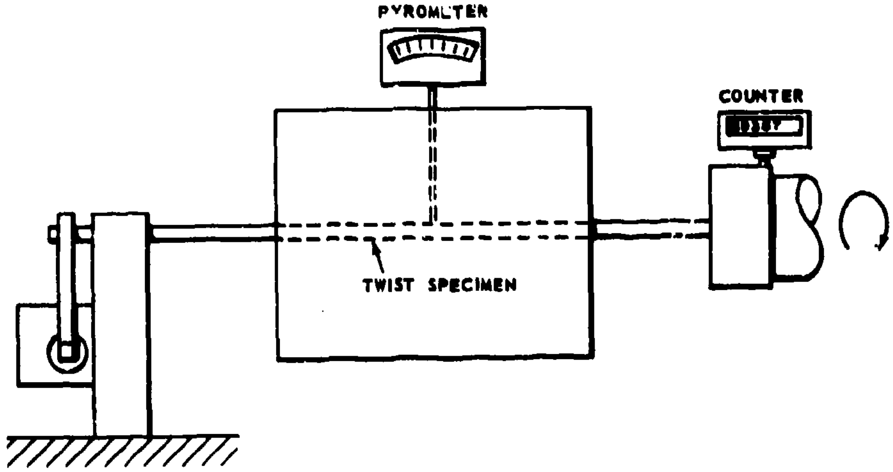


FIGURE 6-1. SCHEMATIC DIAGRAM OF HOT-TWIST TESTING APPARATUS FOR MEASURING FORGEABILITY OF METALS

Upset Test. This test usually consists of upset forging a series of cylindrical billets to various thicknesses, or to the same thickness but with varying billet length-to-diameter ratios. The limit for upset forging without failure by radial or peripheral cracking is considered a measure of forgeability. This test is the one most widely used by the forging industry. It is also useful to some extent for determining billet cleanliness, since segregation, seams, and porosity cause rupturing.

Notched-Bar Upset Test. This test is essentially the same as the upset test except that longitudinal notches are machined into the test bars before upsetting.[6] Proponents of this test argue that the notch causes more severe stress concentrations, representing a more reliable index of the forgeability to be expected in a complex forging die.

Notched-bar upset test samples are prepared by quartering a billet longitudinally into four test bars. This exposes center material along one corner of each test sample. Longitudinal notches are machined into the test samples according to the sketch shown in Figure 6-2. Two notch-root radii (1.0 mm and 0.25 mm) are used on alternate faces. A weld button is often placed on one corner for identifying the center and surface material after forging.

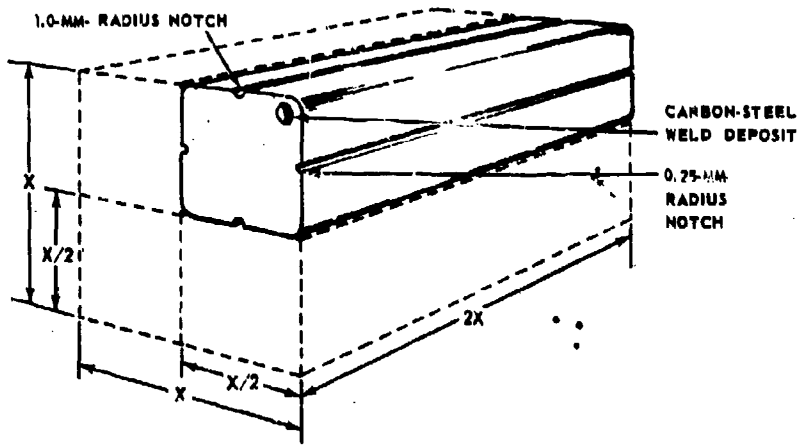


FIGURE 6-2. METHOD OF PREPARING SPECIMENS FOR NOTCHED-BAR UPSET FORGEABILITY TEST

The test samples are heated and upset forged to a height reduction of 4:1 at temperatures prescribed for the material. More than one temperature may be studied, depending on prior experience with the alloy in question. Because of the stress concentrations, the bars are most likely to exhibit rupturing in the notched areas.

The following rating system has been devised for comparing and recording the test data: If no ruptures are observed, a rating of 0 is applied. If the ruptures are small, discontinuous, and scattered, the rating is 1. Higher rating numbers indicate an increasing incidence and depth of ruptures. Figure 6-3 illustrates this suggested numerical rating for four degrees of rupturing.

The usefulness of this test is illustrated in Figure 6-4, which compares rings rolled from heats of Type 403 stainless steel exhibiting forgeability ratings of 0 and 4. The ring rolled from the heat with a rating of 0 was sound, while the ring from the heat with a rating of 4 ruptured extensively.

This test is certainly more sensitive than simple upset testing. Reportedly, unnotched billets from heats having a notched-bar forgeability rating of 3 may be perfectly sound after similar reductions by simple upsetting. In such cases, the simple upset test usually indicates a deceptively higher degree of forgeability than realized in practice. Thus, the notched-bar upset is particularly useful for identifying materials having marginal forgeability.

Hot-Impact Tensile Test. This tension test is conducted in a conventional impact-test machine(7) as shown schematically in Figure 6-5. The test is particularly suitable for determining forgeability of strain-rate-sensitive and precipitation-hardenable alloys.

Rating 1

Rating 2

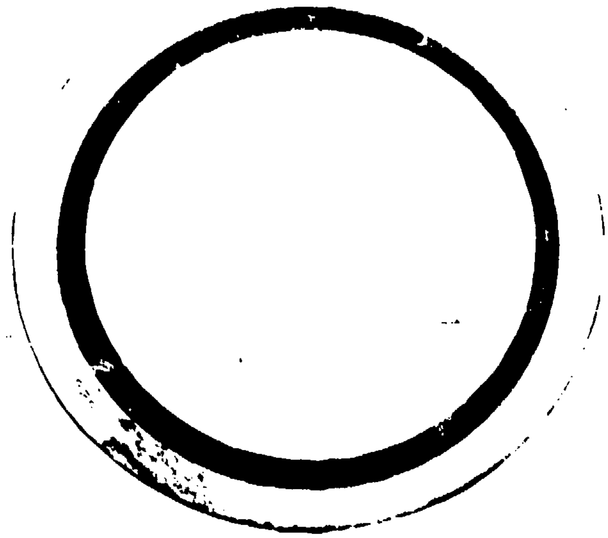
Rating 3

Rating 4

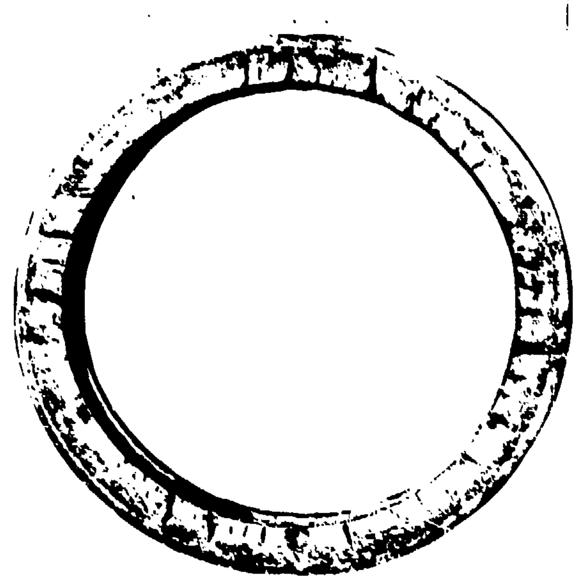
FIGURE 6-3. SUGGESTED RATING SYSTEM FOR NOTCHED-BAR UPSETS THAT EXHIBIT PROGRESSIVELY POORER FORGEABILITY

Rating of 0 would be free of ruptures in notched area.

(Photograph courtesy of The Ladish Company)



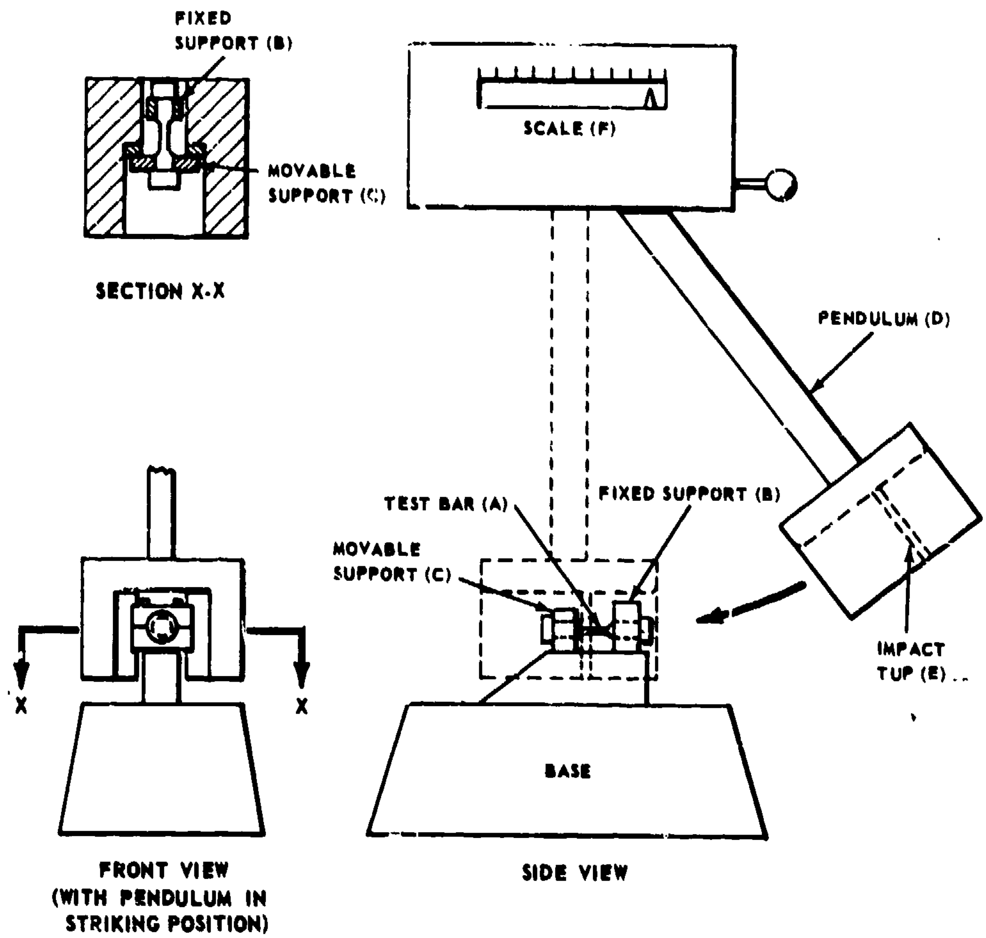
A. Forgeability Rating of Billet = 0



B. Forgeability Rating of Billet = 4

FIGURE 6-4. ROLLED RINGS MADE FROM TWO HEATS OF TYPE 403 STAINLESS STEEL EXHIBITING DIFFERING FORGEABILITY RATINGS IN NOTCHED-BAR UPSETTING TESTS

(Courtesy of The Ladish Company)



SEQUENCE

(1) Heated tension test bar (A) is placed in fixed support (B) and movable support (C)

(2) Pendulum (D) is released and impact tup (E) strikes movable support (C)

(3) Tension test bar breaks and energy (in ft-lb) is read from scale (F)

FIGURE 6-5. SCHEMATIC DIAGRAM OF IMPACT-TEST MACHINE ADAPTED FOR DETERMINING FORGEABILITY

Superalloys, for example, begin to show precipitation reactions at relatively high temperatures. When precipitation takes place, the alloys often demonstrate greatly reduced forgeability along with rapidly increasing resistance to deformation. The hot-impact tensile test gives qualitative data on both properties by measuring, respectively, reduction in area and the foot-pounds of energy. Since the test does not evaluate metal cleanliness or surface quality, it is usually supplemented by a series of upset tests.

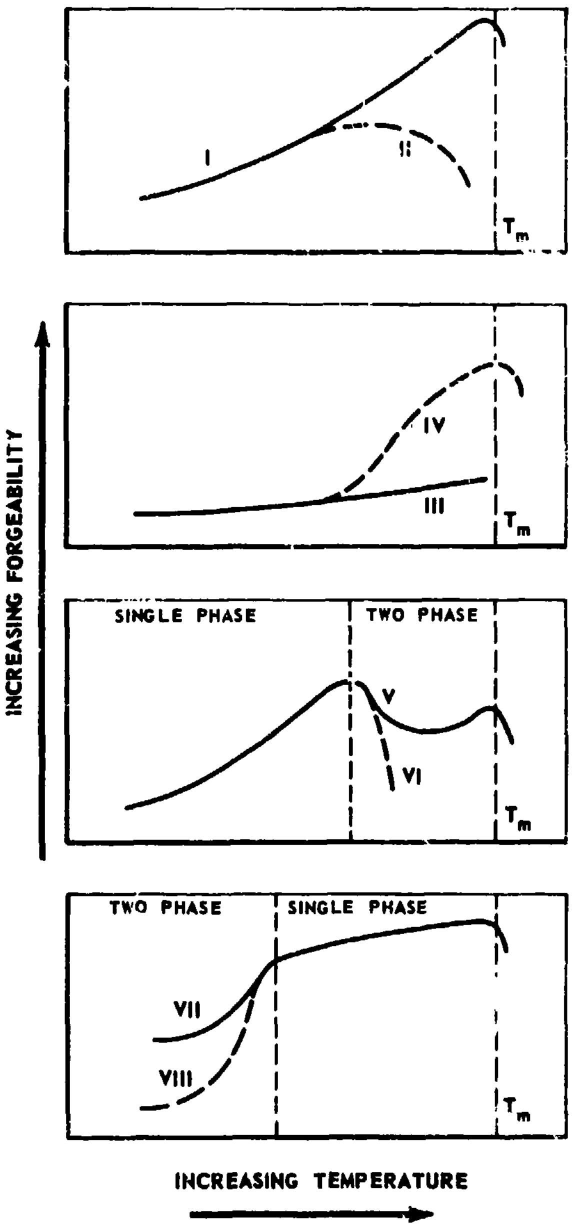
Tension and Compression Tests. These tests are ordinarily performed in conventional tension-test machines which have ram velocities on the order of 1 in./min. Thus, strain rates obtainable are considerably lower than those of ordinary forging equipment. However, tensile-test data (stress, elongation, and reduction in area) obtained at temperatures corresponding to forging temperatures are useful for indicating where behavior such as strain hardening, grain growth, and incipient melting occurs. As the testing temperature is increased, a sharp rise in ductility usually denotes the onset of the hot-working range; a gradual loss of ductility usually indicates grain growth; and a sharp drop of ductility often indicates either incipient melting or the appearance of a second phase. Precipitation-hardenable nickel-base superalloys, for example, ordinarily exhibit a drop in tensile ductility at temperatures where precipitation takes place.

Compression tests would be expected to provide values more useful for estimating forging pressures. For instance, plane-strain compression tests can be made to judge the constraint imposed by undeformed material adjacent to the specimen. Unfortunately, it is desirable that the compression anvils be heated to the testing temperature and, at the same time, be stronger than the material being investigated. Consequently, the compression test is seldom used for evaluating forging characteristics of metals forged much above 1200 F.

Choice of Tests. The proper choice among forgeability tests depends somewhat on the properties of the alloy being tested. Based on known correlations of test values with forging performance, the hot-twist test is particularly useful for evaluating the forgeability of carbon, low-alloy, and stainless steels, which are forged relatively high in their hot-working temperature ranges. However, the test fails to correlate with the forging performance of metals deformed at comparatively low temperatures because work-hardening effects cause the twisting zone to shift during testing. Tension tests, performed either in impact or tensile-test machines, provide data on ductility which agree to some extent with the forgeability of metals at cold-working temperatures. Probably the best combination of tests for forgeability consists of laboratory-scale tests to pinpoint the temperature ranges for optimum forgeability followed by upset (notched or unnotched) tests on full-size billets. The upset tests should be conducted at forging rates comparable to those expected for actual die forging and with a reduction of at least 50 and preferably 75 per cent.

METALLURGICAL FACTORS INFLUENCING FORGEABILITY

Pure metals behave classically during forging in that those having face-centered cubic, body-centered cubic, and close-packed hexagonal crystalline structures generally exhibit decreasing forgeability in that order. Once the pure metals are intentionally or unintentionally alloyed with other elements, however, these classical distinctions



I. Pure Metals and Single-Phase Alloys
(a) Aluminum alloys
(b) Tantalum alloys
(c) Columbium alloys

II. Pure Metals and Single-Phase Alloys Exhibiting Rapid Grain Growth
(a) Beryllium
(b) Magnesium alloys
(c) Tungsten alloys
(d) All-beta titanium alloys

III. Alloys Containing Elements That Form Insoluble Compounds
(a) Resulfurized steel
(b) Stainless steel containing selenium

IV. Alloys Containing Elements That Form Soluble Compounds
(a) Molybdenum alloys containing oxides
(b) Stainless steel containing soluble carbides or nitrides

V. Alloys Forming Ductile Second Phase on Heating
(a) High-chromium stainless steels

VI. Alloys Forming Low-Melting Second Phase on Heating
(a) Iron containing sulfur
(b) Magnesium alloys containing zinc

VII. Alloys Forming Ductile Second Phase on Cooling
(a) Carbon and low-alloy steels
(b) Alpha-beta and alpha-titanium alloys

VIII. Alloys Forming Brittle Second Phase on Cooling
(a) Superalloys
(b) Precipitation-hardenable stainless steels.

T_m = melting temperature.

FIGURE 6-6. EIGHT TYPICAL FORGEABILITY BEHAVIORS EXHIBITED BY DIFFERENT ALLOY SYSTEMS

attributed to atomic structure do not always prevail because of the many other interrelated variables. The most important metallurgical variables are:

(1) Composition and purity
(2) Number of phases present
(3) Grain size.

In general, the forgeabilities of metals and alloys increase with increasing temperature. However, there are eight distinct forgeability behaviors exhibited by the various alloy systems as they are affected by these three variables, with increasing temperature. These behaviors are exemplified by the curves shown in Figure 6-6.

Pure metals and single-phase alloys exhibit increasing forgeability with increasing temperature (Type I). However, as temperature increases, grain growth occurs and some of the pure metals and single-phase alloys exhibit reduced forgeability with increasing grain size (Type II). The adverse affect of increasing grain size on forgeability is particularly significant when brittle compounds are formed on the grain boundaries. Figure 6-7 represents the typical influence of grain size and temperature on the forgeability of solid-solution alloys and alloys containing elements that form insoluble compounds. Metals containing fine grains have much larger grain-boundary surface area than those with coarse grains. Hence, in metals that contain insoluble compounds, a fine grain size produces a lower concentration of the compounds at the grain boundaries and forgeability is far better. Molybdenum and tungsten alloys show this grain-size dependency more than most metals because they have extremely low solubilities for oxygen, nitrogen, and other elements that form brittle compounds. Ordinarily, solid-solution alloys show only slightly reduced forgeability with increasing grain size.

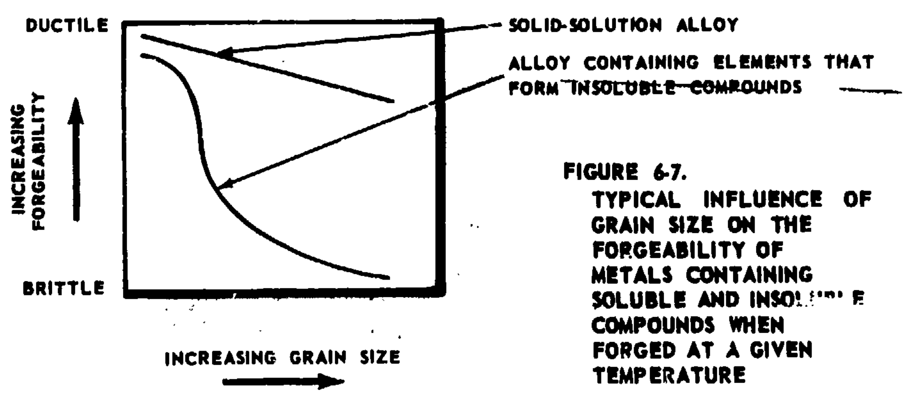


FIGURE 6-7. TYPICAL INFLUENCE OF GRAIN SIZE ON THE FORGEABILITY OF METALS CONTAINING SOLUBLE AND INSOLUBLE COMPOUNDS WHEN FORGED AT A GIVEN TEMPERATURE

Alloys containing elements that form insoluble compounds* (Type III) exhibit brittle behavior regardless of forging temperature. However, if the compounds dissolve with increasing temperature (Type IV), forgeability is improved and, once complete solution takes place, the alloys exhibit forgeability behavior much like that of the pure metals and single-phase alloys.

*Distinction is made between compounds and phases in this discussion. Although sometimes called intermediate phases, compounds are considered as having distinctive stoichiometric ratios of elements.

Thus, the interaction of such factors as grain growth and the formation of insoluble compounds complicates forgeability behavior. For example, an effort to improve the forgeability of an alloy by increasing the forging temperature may actually lead to reduced forgeability because of grain growth and the accompanying brittle grain-boundary condition.

The behaviors of Types V through VIII illustrate how the forgeabilities of alloys are influenced by second phases* of different characteristics. Two-phase alloys are generally less forgeable than are single-phase alloys. This is especially true if the second phase is present in small amounts. In hot-twist studies on chromium and chromium-nickel stainless steels, F. K. Bloom et al.(3), found that hot workability dropped rapidly as the second phase, delta ferrite, began to form upon heating the otherwise austenitic alloys. The hot workability was slowly restored when the amount of delta ferrite increased in proportions above about 15 per cent of the microstructure. Alloys containing totally ferritic structures generally showed better hot-twist ductility than did those containing all austenite. Figure 6-8 depicts how a totally austenitic steel will behave if it transforms entirely to ferrite upon heating. **

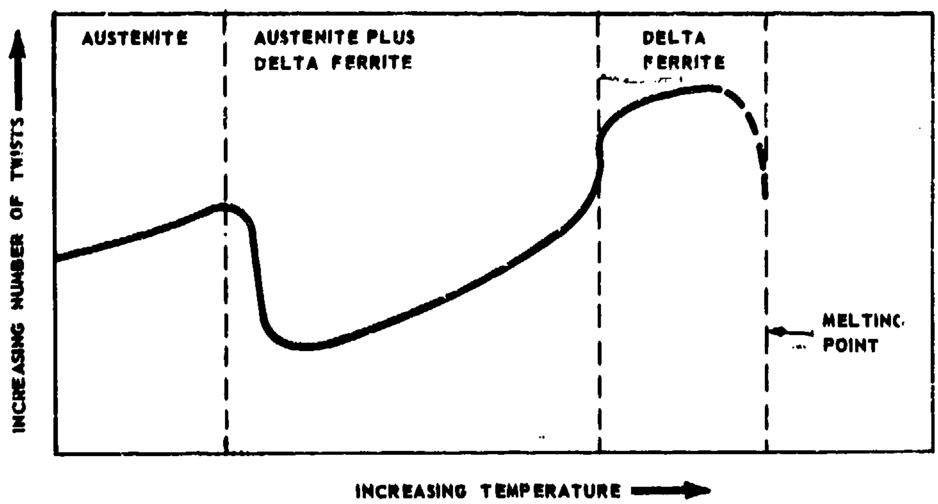


FIGURE 6-8. INFLUENCE OF DELTA FERRITE ON THE TWIST DUCTILITY OF HIGH-CHROMIUM STAINLESS STEELS

It will be noticed that the forgeability of these steels is markedly reduced when either of the two phases is present in small proportions. Yet, both phases are individually ductile.

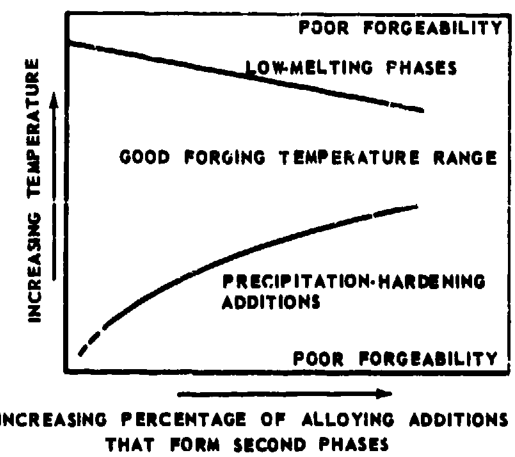
* Distinction is made between compounds and phases in this discussion. Although sometimes called intermediate phases, compounds are considered as having distinctive stoichiometric ratios of elements.

** None of the individual compositions studied by Bloom, et al., covered the full range of behavior indicated; but the authors selected several compositions that contained various proportions of the two phases. In this way they studied the range of microstructures from totally austenitic to totally ferritic.

When a second phase is considerably weaker than the matrix or when the second phase is brittle, forgeability decreases rapidly with increasing amounts of the second phase. Second phases with low melting points have a similar effect. When the melting points of such phases are exceeded during forging, the alloys rupture during deformation. This latter condition is termed "hot shortness", and the temperature at which it occurs is called the "incipient" melting temperature.

Another influence of a second phase is found in alloys which form second phases on cooling from forging temperatures. Most superalloys for example, contain precipitation-hardening additions that cause precipitation to occur in the matrix phase if forging temperatures are below the solid-solution temperatures. This effect becomes increasingly pronounced with increasing amounts of alloying elements that form second phases. Such alloys exhibit minimum as well as maximum forging temperatures, as shown in Figure 6-9. The Carpenter Steel Company(7) has conducted some notable studies of this effect using the hot-impact tensile test. Its work on the iron-base superalloy V-57 showed that ductility was at a maximum between 1700 F and 2000 F, as illustrated in Figure 6-10.

FIGURE 6-9. TYPICAL INFLUENCE OF ALLOYING ELEMENTS THAT FORM SECOND PHASES ON THE RANGE OF TEMPERATURES FOR GOOD FORGEABILITY



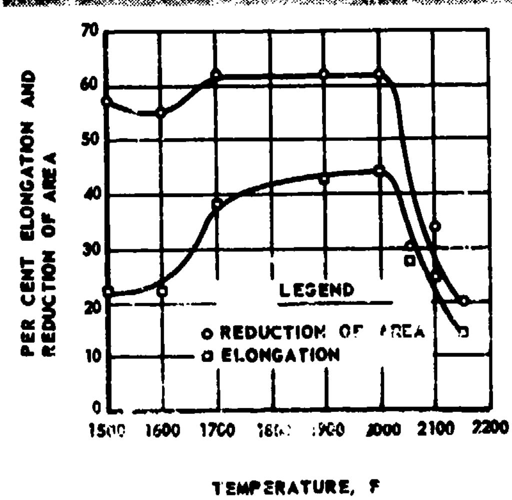


FIGURE 6-10. INFLUENCE OF TEMPERATURE ON THE IMPACT-TENSILE DUCTILITY OF THE IRON-BASE SUPER-ALLOY V-57(7)

(Data courtesy of Carpenter Steel Company)

The quantity, distribution, and shape of second phases and compounds all have a strong influence on forgeability. If these components are concentrated at grain boundaries in the form of distinct globular or isolated inclusions, they are less harmful than if they form a continuous grain-boundary film. The most favorable location of second phases and compounds is within the grains.

The influence of variations in temperature on the forgeability of metals is a subject of considerable study, especially with steels. Contractor and Morgan(8) made an extensive review of the literature on the subject and observed that the hot-twist test seemed to be the best technique for measuring hot workability of steels over the hot-working temperature ranges. The hot-twist test seemed particularly well suited to evaluating forgeability for hot-piercing operations

All investigators concluded that the forgeabilities of metals increase with increasing temperatures until a temperature is reached where either a second phase appears, melting begins, or, in some instances, grain growth is excessive. Practical limits to increasing forging temperatures are imposed by oxidation or contamination reactions that occur when the hot metal is exposed to the atmosphere during forging. These limits are described in later chapters on individual forging-alloy systems.

MECHANICAL FACTORS INFLUENCING FORGEABILITY

In addition to metallurgical factors, strain rate and stress distribution also influence the forgeability of metals. It is widely known, for example, that magnesium and its alloys exhibit considerably better forgeability when deformed in presses than in the faster hammers. Carbon and low-alloy steels, on the other hand, forge as well or better at higher deformation rates. As pointed out earlier, metals exhibit increasing flow stresses with increasing strain rates, and this relative increase becomes more pronounced at higher temperatures. It thus seems logical that ductility should decrease in some proportion to the increase in flow stress. This is essentially true for magnesium and beryllium, but the behavior is not characteristic of carbon and low-alloy steels.

In general, metals exhibiting low ductility at cold-working temperatures show reduced forgeability at increasing strain rates, and metals exhibiting high ductility at cold-working temperatures are not noticeably affected by increasing strain rates. Metals exhibiting these two types of behavior are:

Low Ductility	High Ductility
Beryllium	Silver
Magnesium	Nickel
Zinc	Iron
Zirconium	Copper
Titanium	Aluminum
	Tantalum

It is readily apparent that the least ductile types are those containing the hexagonal close-packed crystalline structures. The forgeabilities of these metals are also the most sensitive to increasing strain rates at forging temperatures.

Metals containing coarse grains are more affected by strain rate than are metals containing fine grains. Early work showed that magnesium could be hammer forged,

providing fine grains were first obtained by press forging.[9] This behavior correlates well with room-temperature test data that show decreasing elongation with increasing grain sizes for AZ80 magnesium alloy.[10]

Grain Size	Elongation, %
ASTM 0-1	3-4
ASTM 5-7	9-12

It is common forging practice to break down coarse-grained, as-cast ingots of alloys such as high-nickel stainless steels, superalloys, and aluminum alloys by press forging rather than by hammer forging because the ingots are less likely to crack during deformation at low strain rates.

When metals are forged rapidly, the work of deformation can cause significant temperature increases that, in turn, can affect forgeability. At very high deformation speeds, the temperature rise may be enough to cause incipient melting or phase change, both of which reduce forgeability. Alloys containing second phases or compounds are more likely to exhibit decreasing forgeability with increasing deformation speeds. On the other hand, single-phase alloys super-saturated with a second phase will usually exhibit better forgeability at higher strain rates. Embrittlement from precipitation of the second phase during deformation may occur at lower strain rates. At the higher rates, precipitation usually takes place after the deformation and the alloy behaves like a normal, single-phase alloy. For these reasons, changes in forgeability attributable solely to strain rate are difficult to isolate from temperature effects.

A significant factor contributing to variations in forgeability is the type of stresses applied to the metal during deformation. During ordinary forging operations, the workpiece is exposed to a combination of compressive, tensile, and shear stresses. Ruptures are normally associated with tensile and shear stresses. As a general rule, therefore, it is important to provide compressive support to those portions of a less forgeable material that are normally exposed to the tensile and shear stresses.

Even simple upset forging can produce a variety of stress conditions that affect forgeability. In studies with copper cylinders, Cook[11] showed that metal deformation during upset forging was nonuniform; the metal in contact with and adjacent to flat-die surfaces was virtually unworked, while the metal toward the specimen centers was deformed more severely than indicated by the simple reduction of height. Studies by Wagner and Boulger[12] and Tomlinson and Stringer[13] show similar nonuniform deformations during upset forging of steel cylinders. Both studies show that, as the cylinders are upset, the unworked regions at the ends (called "dead-metal zones") remain essentially unworked until they meet at the center. In both studies, the dead-metal zone was found to be approximately in the shape of a 120-degree cone. Figure 6-11 shows this effect schematically. Until these cones meet, the dead-metal zones do not begin to deform. Wagner and Boulger's studies were made on low-carbon steel billets with inserts of high-carbon steel rods; the Tomlinson and Stringer studies were made on mild-steel billets with a central hole.

The Tomlinson and Stringer upset studies (Figure 6-11b) showed that the metal at the billet centers flows outward until the dead-metal zones meet in the center. At this point, the metal flow reverses and causes the hole to close and continue to weld as the upset continues. They observed that the dead-metal zones retain essentially the same conical geometry for cylinders varying in height-to-diameter ratios between 1:1 and 2:1.

providing fine grains were first obtained by press forging.[9] This behavior correlates well with room-temperature test data that show decreasing elongation with increasing grain sizes for AZ80 magnesium alloy.[10]

Grain Size	Elongation, %
ASTM 0-1	3-4
ASTM 5-7	9-12

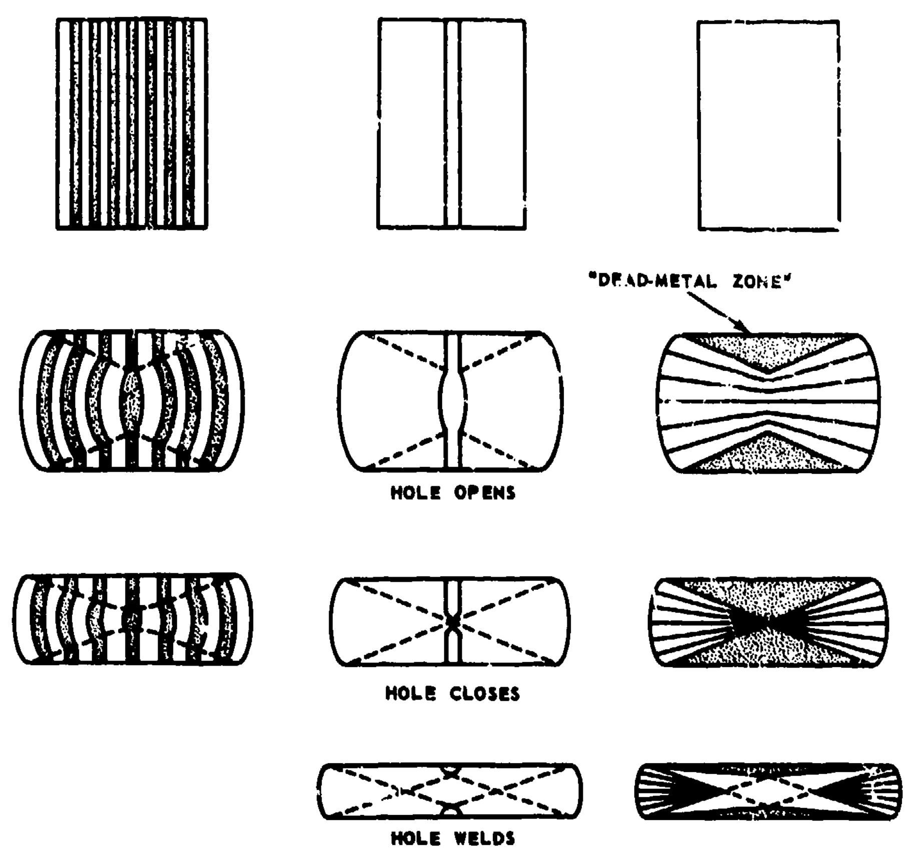
It is common forging practice to break down coarse-grained, as-cast ingots of alloys such as high-nickel stainless steels, superalloys, and aluminum alloys by press forging rather than by hammer forging because the ingots are less likely to crack during deformation at low strain rates.

When metals are forged rapidly, the work of deformation can cause significant temperature increases that, in turn, can affect forgeability. At very high deformation speeds, the temperature rise may be enough to cause incipient melting or phase change, both of which reduce forgeability. Alloys containing second phases or compounds are more likely to exhibit decreasing forgeability with increasing deformation speeds. On the other hand, single-phase alloys super-saturated with a second phase will usually exhibit better forgeability at higher strain rates. Embrittlement from precipitation of the second phase during deformation may occur at lower strain rates. At the higher rates, precipitation usually takes place after the deformation and the alloy behaves like a normal, single-phase alloy. For these reasons, changes in forgeability attributable solely to strain rate are difficult to isolate from temperature effects.

A significant factor contributing to variations in forgeability is the type of stresses applied to the metal during deformation. During ordinary forging operations, the workpiece is exposed to a combination of compressive, tensile, and shear stresses. Ruptures are normally associated with tensile and shear stresses. As a general rule, therefore, it is important to provide compressive support to those portions of a less-forgeable material that are normally exposed to the tensile and shear stresses.

Even simple upset forging can produce a variety of stress conditions that affect forgeability. In studies with copper cylinders, Cook[11] showed that metal deformation during upset forging was nonuniform; the metal in contact with and adjacent to flat-die surfaces was virtually unworked, while the metal toward the specimen centers was deformed more severely than indicated by the simple reduction of height. Studies by Wagner and Boulger[12] and Tomlinson and Stringer[13] show similar nonuniform deformations during upset forging of steel cylinders. Both studies show that, as the cylinders are upset, the unworked regions at the ends (called "dead-metal zones") remain essentially unworked until they meet at the center. In both studies, the dead-metal zone was found to be approximately in the shape of a 120-degree cone. Figure 6-11 shows this effect schematically. Until these cones meet, the dead-metal zones do not begin to deform. Wagner and Boulger's studies were made on low-carbon steel billets with inserts of high-carbon steel rods; the Tomlinson and Stringer studies were made on [illegible] steel billets with a central hole.

The Tomlinson and Stringer upset studies (Figure 6-11b) showed that the metal at the billet centers flows outward until the dead-metal zones meet in the center. At this point, the metal flow reverses and causes the hole to close and continue to weld as the upset continues. They observed that the dead-metal zones retain essentially the same cone geometry for cylinders varying in height-to-diameter ratios between 1:1 and 2:1.



A. Deformation Pattern in Upset Billet With Inserted Grid Rods [Wagner & Boulger[12]]

B. Deformation Pattern in Upset Billet With Center Hole [Tomlinson & Stringer[13]]

C. Distribution of Deformation Based on A and B

- Unworked metal
- Firs[illegible]form
- Last to deform

FIGURE 6-11. DEFORMATION PATTERNS IN CYLINDRICAL BILLETS UPSET FORGED BETWEEN FLAT DIES

These points about upset forging are clear:

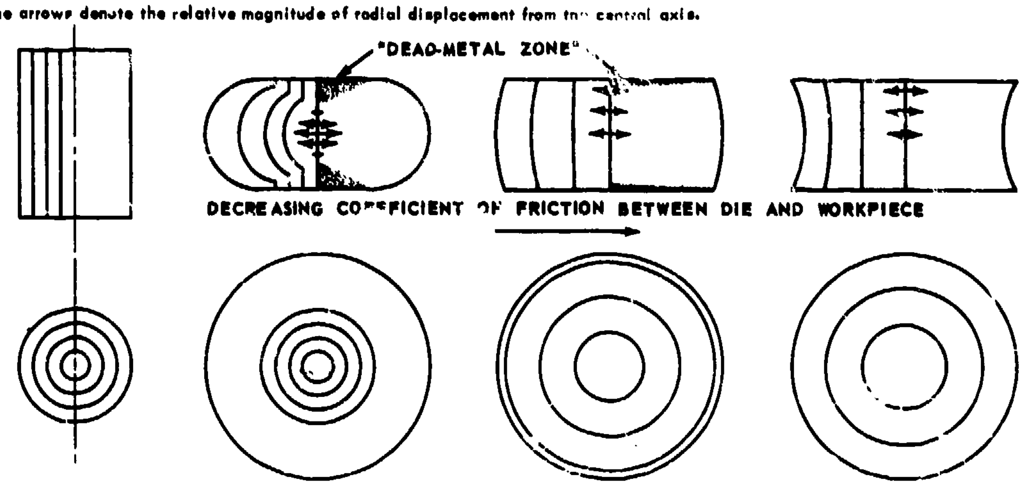
(1) Metal deformation during forging is nonuniform.

(2) Billet centers are exposed to radial tensile stresses during the initial stages of forging.

(3) Stresses at the billet centers become entirely compressive once the dead-metal zones meet. This occurs at a height-to-diameter ratio of approximately 1:2.

When lubrication is applied to flat dies during upset forging, these stress-distribution patterns change considerably. An analysis of work done by three investigators[12,14,15] shows that the shape of upset-forged samples can change with decreasing friction, as illustrated in Figure 6-12. The increasing diameters of the concentric rings show that decreasing the surface friction causes increasing amounts of surface-metal flow. This produces more uniform stress distribution and, in turn, more uniform metal flow (Figure 6-12c). When friction is almost completely eliminated, conditions can be such that the surface metal spreads more than the center material (Figure 6-12d). This occurs frequently when forging metals such as aluminum and magnesium on dies heated to or above the temperature of the workpiece. Thus, an infinite combination of stresses can exist during upset forging depending on (1) the h/D ratio and (2) the amount of surface spread.

The arrows denote the relative magnitude of radial displacement from the central axis.



A. Original Billet With Equal-Spaced Grid Markings

B. Little or No Surface Spread; Large Dead-Metal Zone

C. Surface Spreads Almost as Much as Center; Shallow Dead-Metal Zone

D. Surface Spreads More Than Center; No Dead-Metal Zone

FIGURE 6-12. SCHEMATIC DRAWING SHOWING THE INFLUENCE OF DECREASING FRICTION ON THE DISTRIBUTION OF DEFORMATION IN UPSET FORGINGS

Table 6-1 lists several metals and alloys that are considered less-forgeable materials. They are grouped into categories according to the range of upset reductions obtainable between flat dies. Forgeability during upset forging depends on the nature of the billet to some extent. A weak-centered billet, for example, is best upset forged

with minimum spread to avoid ruptures at the die surface. Conversely, a billet containing surface and subsurface defects is best upset forged with maximum spread to provide maximum compressive stress at the billet sides. Relatively brittle materials are best forged with an h/D ratio of less than 1:1 between nonlubricated dies. Center material is then exposed to compressive stresses at the outset of deformation and therefore is less likely to rupture. This observation corresponds well with the practices used to forge metals exhibiting poor forgeability (e.g., free-machining steels, two-phase alloys, and cast ingots). Such metals are usually upset forged with minimum h/D ratios. For example, one company(9) reports the use of 8-inch-diameter billet stock cut to lengths of 2 to 3 inches for forging 9- to 10-inch-diameter disks from Types 321 and 347 stainless steels.

TABLE 6-1. EXAMPLES OF LESS-FORGEABLE METALS AND ALLOYS

Metal or Alloy	Condition	Approximate Range of Upset Reductions Obtainable Without Fracture Between Flat Dies: 0-10 Per Cent	10-25 Per Cent	25-50 Per Cent
Magnesium				
AZ80	As cast		X	
ZK60	As cast			X
Columbium				
Cb-1Zr	Arc cast			X
Nickel				
Hastelloy W	Air melted			X
Inco 901	Air melted		X	
	Arc cast		X	
Waspaloy	Arc cast		X	
	Wrought			X
René 41	Arc cast		X	
	Wrought			X
Molybdenum	Arc cast	X		
	Arc cast and extruded			X
Tungsten	Arc cast	X		
	Arc cast and extruded			X
	Pressed and sintered			X
Chromium	Arc cast	X		
Beryllium	Arc cast	X		
	Arc cast and extruded		X	
	Hot pressed		X	

Methods of Improving Forgeability

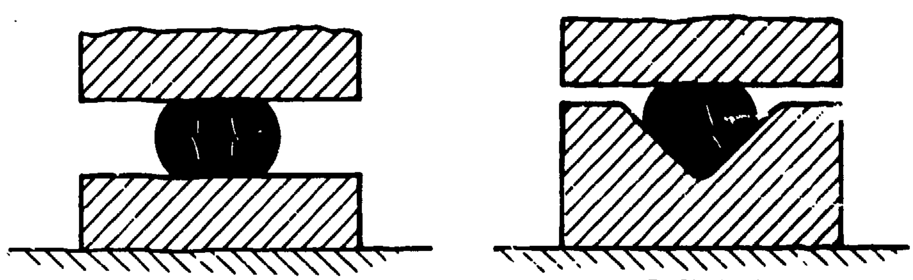
It has been well established both by theory and in practice that metals comparatively brittle when forged without lateral support between flat dies can be deformed successfully when supported completely by compressive stresses as in extrusion. There are several techniques for affording this necessary support in forging. Some involve the use of specially designed dies. Others utilize more ductile materials in the form of thick-walled cans, rings, sleeves, etc., for containment of the workpiece, sometimes

in combination with specially designed dies. When considering die-forging design with these latter techniques, it is important to adapt the shape to accommodate the added volume of expendable material. Because simultaneous metal flow of two dissimilar metals is difficult to predict accurately, the designs for the less-forgeable materials are generally conservative and represent a minimum complexity in shape.

Vee-Flat Dies. An example of a method for improving forgeability by changing the stress distribution during forging is exemplified by the "Vee-flat" die arrangement used in the conversion of weak-centered ingots. As shown in Figure 6-13, flat dies produce maximum stress at the weak billet centers while the Vee-flat dies shift the maximum stress to the mid-radius positions. Ingots containing center segregation, center porosity, or coarse grains will usually exhibit considerably improved forgeability when forged in the Vee-flat dies.

Multiple-Stage Upsetting Dies. Alloys that develop ruptures after small upset reductions, e.g., 20-25 per cent, may be upset forged in a series of cylindrical impression dies having progressively larger diameters, each permitting 20-25 per cent reduction. During this sequence, the bulging common to ordinary upset forging between flat dies is prevented and the billet is upset to cylinders with progressively larger diameters and shorter lengths. It does not take many stages, however, before the cost of such an approach becomes expensive.

Essentially, the same principle is employed when upset forging between opposed conical dies of the type illustrated in Figure 6-14. The tapered cavities of such dies provide progressively increasing lateral support. The accumulated small reductions imparted by these techniques usually provide enough deformation to permit recrystallization and grain refinement. When this occurs, forgeability often improves progressively.



Arrows denote direction and location of maximum stress.

FIGURE 6-13. PRINCIPLE OF VEE-FLAT FORGING DIE SHOWING HOW MAXIMUM STRESSES ARE SHIFTED AWAY FROM THE CENTER OF A FORGING BILLET

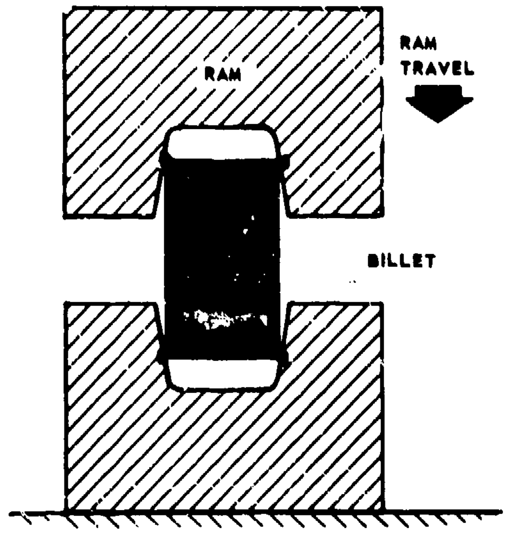


FIGURE 6-14. SKETCH OF POT DIES DESIGNED TO PROVIDE COMPRESSIVE STRESSES ON BILLETS DURING UPSET FORGING

Back-Extrusion Dies. The back-extrusion process is a useful method for forging under essentially compressive stress. In this case, the billet is placed in a container matching the billet's dimensions and shape. A punch is forced into the billet and the metal extrudes into the annular space between the container and the punch. The only portion of the billet that is unsupported by the tools is at the top edges of the part. For extremely brittle materials, additional support can be provided in this region by using an expendable ring of back-up material in the annular space. These back-extrusion methods are illustrated in Figure 6-15. The use of an expendable back-up material requires a double-action forging press. This is necessary to provide a controlled force on the pressure ring that will permit the expendable material to extrude between the pressure ring and the container as the punch advances. For obvious reasons, these methods are suitable only to cuplike shapes that are producible by back-extrusion techniques.

Thick-Walled Sleeves and Cans. About the most versatile method for providing compressive support to forging billets is the use of thick-walled sleeves and cans made from expendable material. However, the process is also probably the most expensive because the canning material must be machined to fit the billet. Ordinarily, the material used for canning has greater plasticity than the billet and thus it flows differently during forging. Figure 6-16 shows the typical distribution of metal forged in three configurations. The more plastic canning material generally gathers in the recesses of contour dies.

In summary, comparatively brittle materials can be forged using one or more of these special techniques. For example, beryllium forging studies at three companies(16, 17,18) have demonstrated that hot-pressed, arc-cast, and arc-cast and extruded beryllium can be forged by methods which offer lateral constraint. For obvious reasons, however, the designs are limited to simple, generally axisymmetrical shapes.

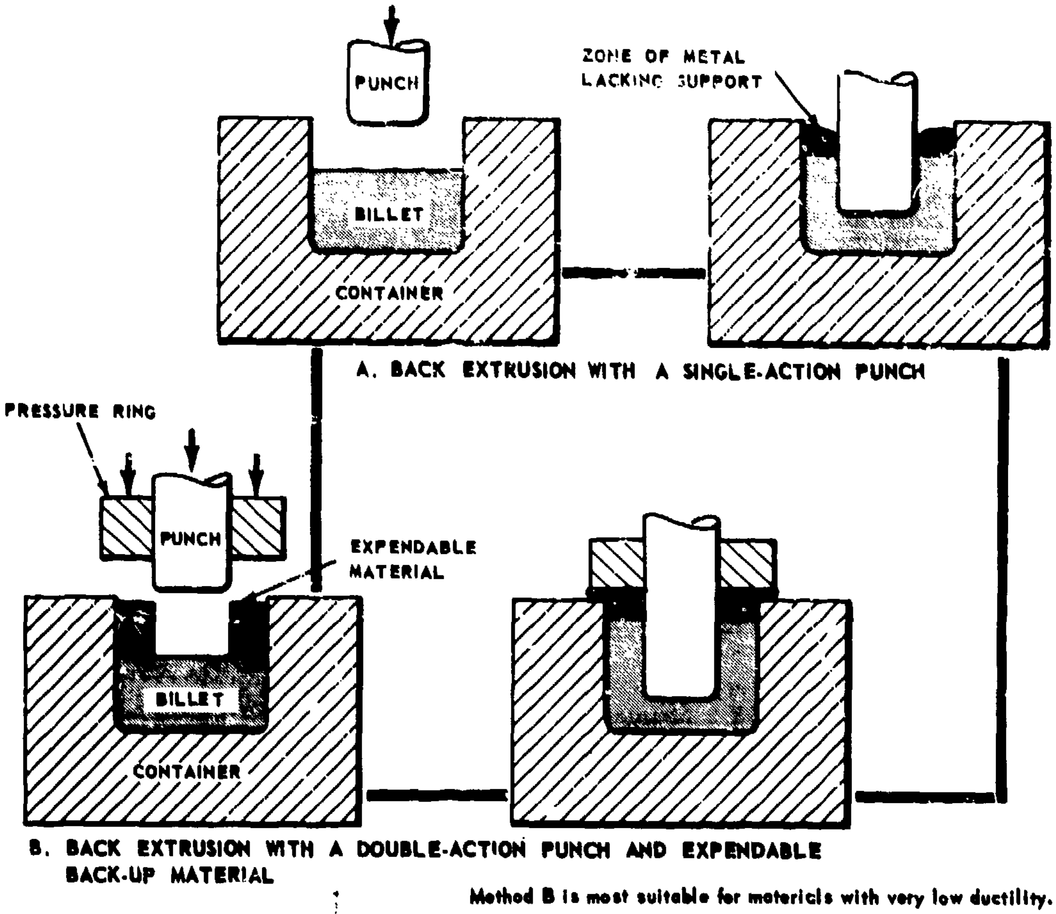


FIGURE 6-15. BACK-EXTRUSION METHODS FOR FORGING UNDER COMPRESSIVE LOADING ON THE WORKPIECE

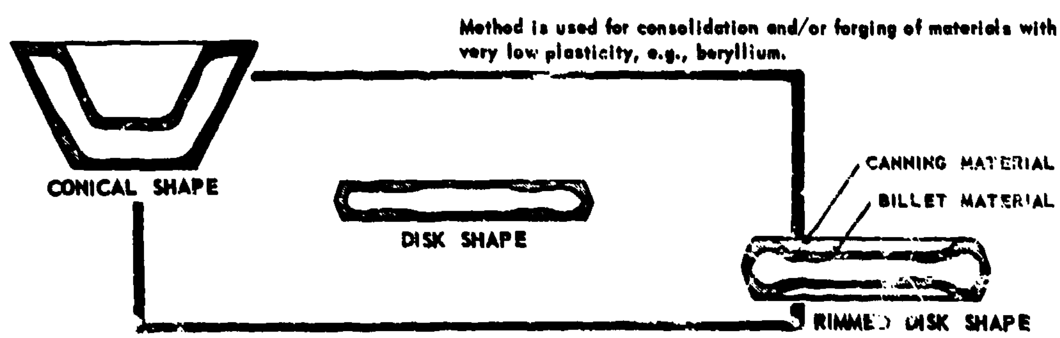


FIGURE 6-16. TYPICAL DISTRIBUTION OF METAL WHEN FORGING MATERIALS ARE ENCAPSULATED IN THICK-WALLED CANS

Often these techniques can be used to insure recrystallization and grain refinement during subsequent annealing. In the majority of cases, this improves forgeability enough to permit subsequent forging in more conventional designs.

PREDICTING FORGING BEHAVIOR

As a general rule, the differences in forgeability among alloy groups are generally greater than among alloys within any one group. Within the iron-base and nickel-base alloy systems, however, the alloys vary from highly forgeable to those which are comparatively brittle even under the most carefully controlled conditions. Thus, in attempting to rank the commonly forged materials in terms of ease of forging, it is necessary to classify the materials both by alloy system and by alloy group within the alloy system. In order of increasing forging difficulty, such a classification would be:

Aluminum alloys
Magnesium alloys
Copper alloys
Carbon and low-alloy steels
Martensitic stainless steels
Maraging steels
Austenitic stainless steels
Nickel alloys
Semiaustenitic PH stainless steels
Titanium alloys
Iron-base superalloys
Cobalt-base superalloys
Columbium alloys
Tantalum alloys
Molybdenum alloys
Nickel-base superalloys
Tungsten alloys
Beryllium

Data on the forging behavior and other characteristics for most of these alloy groups are given in later chapters. However, it is worthwhile to review here the steps necessary for making predictions about forging behavior. The first step is to define the important physical characteristics that influence plastic behavior. Briefly, these are:

(1) Crystalline Structure – The three basic metallic crystalline structures, close-packed hexagonal, face-centered cubic, and body-centered cubic, are normally associated with increasing workability, in that order.

(2) Phase Changes – Transformations from one crystalline structure to a more ductile structure during heating have a pronounced effect on the plasticity of a metal and, thus, are important in selecting working temperatures.

(3) Number of Phases – Since alloys containing two phases are generally less forgeable than those containing one phase, forging as a general rule should be done at a temperature sufficiently high to dissolve the second phase completely.

(4) Solidus Temperature — The lowest temperature at which an alloy begins to melt defines the upper limiting temperature for forging.

(5) Recrystallization Temperature — The recrystallization temperature of a metal defines the limits of hot and cold working.

(6) Grain Size — The ductility and, hence, the plasticity of an alloy improve with decreasing grain size.

(7) Chemical Reactions — Knowledge of surface reactions of a metal when heated in certain environments is important since many reaction products have an effect on such factors as forgeability, friction, and final forging quality.

Having such information available, first predictions of forging behavior can be made. The following examples are typical approaches for two metal alloys, A and B. Alloy A has a face-centered cubic crystalline structure and contains fine grains. The alloy retains a single-phase cubic structure from absolute zero to melting, which starts at 5300 F. The alloy recrystallizes when heated in the neighborhood of 2200 F, and grain growth occurs above this temperature. When heated to 1000 F, the alloy begins to oxidize. At higher temperatures, the alloy absorbs both oxygen and nitrogen, forming a hard, contaminated surface.

Alloy B has a two-phase structure at room temperature. One phase is close-packed hexagonal, the other is body-centered cubic. Upon heating, the hexagonal structure transforms continuously to the cubic structure, until the alloy assumes the cubic structure completely at 1800 F. The presence of the hexagonal phase retards grain growth of the cubic phase during heating. Above 1800 F, the single-phase cubic structure exhibits rapid grain growth. The cubic structure is retained until melting begins at 2900 F. Similar to Alloy A, Alloy B oxidizes slowly until heated above 1000 F, where both oxidation and contamination become increasingly rapid.

From these brief descriptions of Alloys A and B*, it is possible to make the following predictions about forging behavior:

(1) Alloy A will require increasing forging pressure with increasing reductions when forged at temperatures up to 2200 F (cold-working behavior).

(2) Alloys A and B will exhibit hot-working behaviors at forging temperatures above 2200 F and 1800 F, respectively.

(3) To prevent oxidation and contamination, both alloys require protection by suitable atmospheres or coatings if heated above 1000 F.

(4) Alloy B should be forged at temperatures in the vicinity of 1800 F. Higher temperatures would promote grain growth; lower temperatures would reduce forgeability.

(5) Alloy A should be ductile at low temperatures; thus, forging temperatures of 1000 F or lower are suitable. Use of forging temperatures above 1000 F would provide little significant advantage unless the temperature exceeded 2200 F, where the alloy would exhibit hot-working behavior.

* Alloys fitting the descriptions of Alloys A and B are Ta-30Cb-7.5V and Ti-6Al-4V, respectively.

This knowledge of expected forging behavior of the alloys in itself is not sufficient to establish a suitable forging practice. In addition, it is necessary to obtain at least the following types of data on the alloys:

(1) Flow Stress and/or Upsetting Pressure -- Measurements of flow stress and/or upsetting pressure are useful for estimating forging pressures and equipment requirements for a given alloy. The upsetting pressure is a better measure of forging-pressure requirements because it takes into account the effects of friction.

(2) Tensile Strength and Elongation – Tensile properties at elevated temperatures usually are more readily available than data on flow stress or upsetting pressure and can be used as a guide to the selection of forging conditions. Comparisons of yield strength provide a measure of the relative forging-pressure requirements. Similarly, data on tensile elongation and reduction of area at elevated temperatures are useful for estimating forging-temperature ranges.

(3) Plastic Range – It is important to know the temperature range where grain growth becomes excessive because of its effect on plasticity. In general, this temperature range becomes the upper limit for forging. The lower limit is determined by phase changes, embrittlement, or an impractically high forging pressure.

(4) Strain-Rate Sensitivity – All metals exhibit an increase in flow stress with increasing strain rate, but do not necessarily exhibit reduced ductility. Data on the effect of strain rate obtained by conducting comparable upset forgeability tests in both hammers and presses are useful in selecting forging equipment.

It should be emphasized, in addition, that, no matter how effective a method seems to be for selecting conditions for forging, it is impossible to expect a consistent response to forging if the starting material is not uniform. Segregation, porosity, and heterogeneity all contribute to forgeability variations. Common practice in forge shops, therefore, consists of conducting preliminary upset forgeability studies on many of the incoming forging billets to assure a reasonable degree of uniformity.

REFERENCES

(1) Ihrig, H. K., "The Effects of Various Elements on the Hot Workability of Steel", Trans. AIME, 167, 749-777 (1946).

(2) Ihrig, H. K., "A Quantitative Hot Workability Test For Metals", Iron Age, 153 (16), 86-89 (April 20, 1944).

(3) Bloom, F. K., Clarke, W. C., Jr., and Jennings, P. A., "Relation of Structure of Stainless Steel to Hot Ductility", Metal Progress, 59, 250-256 (February, 1951).

(4) Clark, C. L., and Russ, J. J., "A Laboratory Evaluation of the Hot-Working Characteristics of Metals", Trans. AIME, 167, 736-748 (1946).

(5) Tout, C. S., Jr., and Banning, L. H., "A Rapid Twist Test For Determining Hot Forming Temperatures of Steels", Report of Investigations 5928, U.S. Dept. of Interior, Bureau of Mines (1962).

(6) Private Communications from The Ladish Company, Cudahy, Wisconsin.

(7) Private Communication from A. R. Walsh of the Carpenter Steel Company, Reading, Pennsylvania.

(8) Contractor, G. P., and Morgan, W. A., "Forgeability of Steels - A Critical Survey of the Literature: Part I", Metal Treatment and Drop Forging, 26 (161), 65-71 (February, 1959). "Part II", ibid., 107-114 (March, 1959).

(9) Private Communication with A. L. Tribe of Wyman-Gordon Company.

(10) Roberts, C. S., Magnesium and Its Alloys, John Wiley and Sons, Ltd., New York (1960).

(11) Cook, Maurice, "Resistance of Copper and Copper Alloys to Homogeneous Deformation in Compression", J. of the Institute of Metals, 71, 371-390 (1945).

(12) Wagner, H., and Boulger, F. W., "A Study of Possible Methods for Improving Forging and Extruding Processes for Ferrous and Nonferrous Materials", Air Force Contract 33(600)-26272, Battelle Memorial Institute (June 30, 1957).

(13) Tomlinson, A., and Stringer, J. D., "The Closing of Internal Cavities in Forgings by Upsetting", J. of the Iron and Steel Institute, 188, 209-217 (March, 1958).

(14) Tardif, H. P., "Deformation Studies in Metalworking Processes", Steel Processing and Conversion, 43, 626-632 (November, 1957).

(15) Tholander, Erik, "Forging Research in Sweden", Metal Treatment and Drop Forging, 28, 89-94 (March, 1961).

(16) Denny, J. P., and Rubenstein, H. S., "The Forging of Jacketed and Bare Beryllium", paper given at Conference on the Metallurgy of Beryllium held October 16-18, 1961, in London by the Institute of Metals.

(17) Hayes, A. F., and Yoblin, J. A., "New Concepts for Fabricating Beryllium Through Advanced Forging and Powder Metallurgy Techniques", paper given at Conference on the Metallurgy of Beryllium held October 16-18, 1961, in London by the Institute of Metals.

(18) Cieslicki, M. E., "Beryllium Forging", paper given at Conference on The Metallurgy of Beryllium held October 16-18, 1961, in London by The Institute of Metals.

CHAPTER 7

ALUMINUM FORGING ALLOYS

Crystalline Structure:	Face-centered cubic
Phase Changes:	None
Number of Phases:	Generally one phase at forging temperatures
Liquidus Temperature Range:	1180 to 1215 F
Solidus Temperature Range:	890 to 1195 F
Reactions When Heated in Air:	Oxidation (not continuous)

ALLOYS AND FORMS AVAILABLE

The compositions of seven aluminum alloys most frequently specified in the form of die forgings are given in Table 7-1. Included are data useful for predicting forging behavior of each alloy. Data for unalloyed aluminum (Grade 1100) are included for comparison purposes.

Billet stock is produced from continuously cast ingots that are normally converted by press forging to square bars and then rolled to rounds or plate. Large billets (over 12-inch-square cross section) are usually furnished as square bars.

FORGING BEHAVIOR

Typical stress-strain curves for 1100 aluminum at various temperatures and at a constant strain rate are given in Figure 7-1. Figure 7-2 shows how the flow stress at a constant reduction increases with increasing strain rate at various temperatures. When these graphs are combined they may be used to form a family of stress-strain curves for each temperature, depending on the strain rate. Four sets of such stress-strain curves are given in Figure 7-3 showing how the stress shifts upward with increasing strain rate. The stress-strain curves at 840 F actually overlap the curves obtained at 1020 F. At these temperatures, a 40-fold increase in strain rate is essentially equal to a 180 F decrease in temperature. Kent[5] studied the influence of temperature on the reduction obtainable on 1100 aluminum when forged under constant drop-weight conditions. The findings of this work are presented in Figure 7-4 which shows quantitatively how the upset reduction at constant input energy increases with increasing temperature.

TABLE 7-1. COMPOSITIONS AND FORGING CHARACTERISTICS OF SEVERAL ALUMINUM ALLOYS[1,2,3]

Alloy Designation	1100	2014	2024	2025	2218	6061	7075	7079
Forging Specifications		QQ-A-367 AMS [illegible] AMS 4135		QQ-A-367 AMS 4130	QQ-A-367 AMS 4142	QQ-A-367 AMS 4150	QQ-A-367 AMS 4139	QQ-A-367 AMS 4138
Nominal Composition, %	99.00 min Al	4.4 Cu 0.8 Si 0.8 Mn 0.4 Mg	4.5 Cu 0.6 Si 1.5 Mg	4.5 Cu 0.8 Si 0.8 Mn	4.0 Cu 2.0 Ni 1.5 Mg	0.25 Cu 0.6 Si 1.0 Mg 0.25 Cr	1.5 Cu 2.9 Mg 0.3 Cr 5.5 Zn	0.6 Cu 3.3 Mg 0.2 Cr 0.2 Mn 4.3 Zn
Solidus, F	1195	950	935	970	940	1080	890	900
Recrystallization Temp, F	500	500	500	450	450	[illegible]	400	400
Forging-Range Temp, F	650-1050	700-900	700-900	800-920	700-900	700-950	700-850	700-850
Normal Forging Temp, F								
Presses	850	850	850	890	850	880	750	750
Hammers	800	820	800	850	800	850	725	725
Relative Forgeability								
Presses	Excellent	Excellent	Excellent	Excellent	Excellent	Excellent	Excellent	Excellent
Hammers	Excellent	Excellent	Excellent	Excellent	Good	Excellent	Fair	Fair
High-Energy-Rate Machines	Excellent	Good	--	--	--	Good	Fair	--

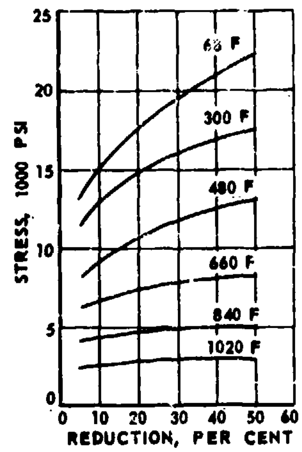


FIGURE 7-1. EFFECT OF TEMPERATURE ON THE STRESS-STRAIN CURVES FOR ANNEALED 1100 ALUMINUM AT A CONSTANT STRAIN RATE

Strain Rate: 4.4 in./in./sec
Sample Size: 1-inch-diam x 0.72 inch

Data after Alder and Phillips[4]

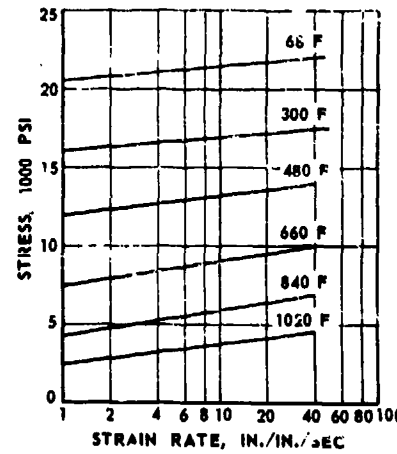


FIGURE 7-2. EFFECT OF STRAIN RATE ON THE STRESS REQUIRED TO COMPRESS ANNEALED 1100 ALUMINUM TO A 40 PER CENT REDUCTION AT VARIOUS TEMPERATURES

Data after Alder and Phillips[4]

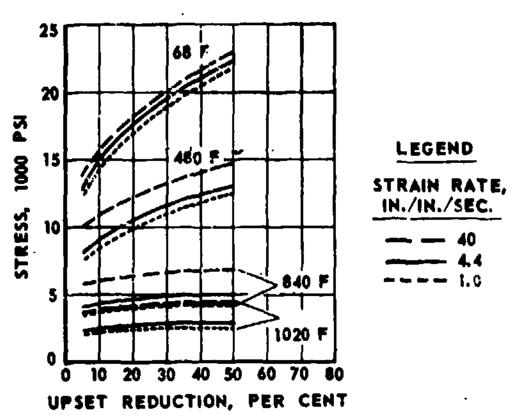


FIGURE 7-3. INFLUENCE OF STRAIN RATE ON THE STRESS-STRAIN CURVES FOR ANNEALED 1100 ALUMINUM AT VARIOUS TEMPERATURES

Data after Alder and Phillips[4]

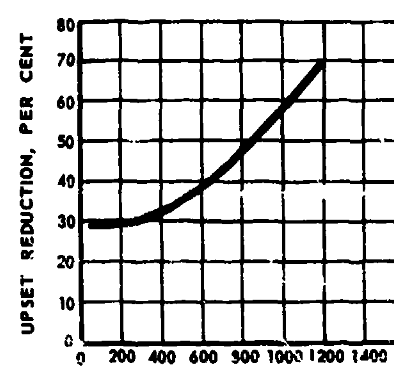


FIGURE 7-4. EFFECT OF TEMPERATURE ON AMOUNT OF UPSET AT CONSTANT DROP FORGING ENERGY FOR 1100 ALUMINUM

Energy: 50 ft-lb
Sample: 1/2-in. diam x 1/2 in.

Data after Kent[5]

Tensile strengths and elongations of several aluminum alloys at various temperatures are presented in Figures 7-5 and 7-6, respectively. Table 7-2 presents elevated temperature yield strengths of two alloys when tested at two strain rates, and shows that the yield strength increases with increasing strain rate. This is similar to the strain-rate effect described earlier for 1100 aluminum.

TABLE 7-2. INFLUENCE OF STRAIN RATE ON THE ELEVATED-TEMPERATURE YIELD STRENGTHS OF TWO ALUMINUM ALLOYS[6]

Alloy	Strain Rate, in./in./sec	Yield Strength, psi, at Temperatures Indicated[a] 500 F	600 F	700 F
2014	0.01	35,000	25,000	5,000
	1.0	42,000	30,000	--
7075	0.01	30,000	20,000	5,000
	1.0	38,000	24,000	--

(a) Data for both alloys are for the solution-treated-and-aged condition (T6).

Neither of the above forms of data indicate the actual forging pressures because the values were determined at strain rates considerably lower than that of conventional forging equipment. However, the data provide useful information about relative forging pressure and forgeability when comparing the alloys with each other. For example, the 7075 alloy which is relatively difficult to forge has nearly ten times the strength of 1100 aluminum at 700 F.

The elongation of all the alloys rises sharply at temperatures corresponding approximately with the onset of recrystallization. With increasing temperatures, the elongation values reach a maximum, then begin to drop - first gradually as grain growth occurs, then sharply when melting begins. For these reasons, elevated-temperature elongation values serve as a guide to selecting forging temperature by indirectly identifying the temperatures for recrystallization, grain growth, and melting.

Figure 7-7 compares elevated-temperature strength properties with actual forging-pressure data for 1100 and 6061. The forging-pressure curves are given for increasing levels of reduction. As expected, the forging pressures are higher, but are in approximately the same proportion as the tensile strengths. On the basis of the curves shown for 6061, the alloy changes from cold-working behavior to hot-working behavior gradually between 750 F and 900 F under press-forging conditions.

INFLUENCE OF FORGING VARIABLES ON MECHANICAL PROPERTIES

Mechanical properties of aluminum-alloy forgings are developed primarily through a duplex heat treatment. For most alloys this consists of solution heat treating at a temperature just below the solidus temperature, water quenching, then aging between 250 and 350 F for several hours. The heat treatment of specific aluminum alloys is well covered in the 8th Edition of the ASM Metals Handbook, Vol 1[3].

Heat treatable aluminum alloys are generally strengthened slightly by forging in the hot-working range. This is due, in part, to the accompanying reduction of grain size

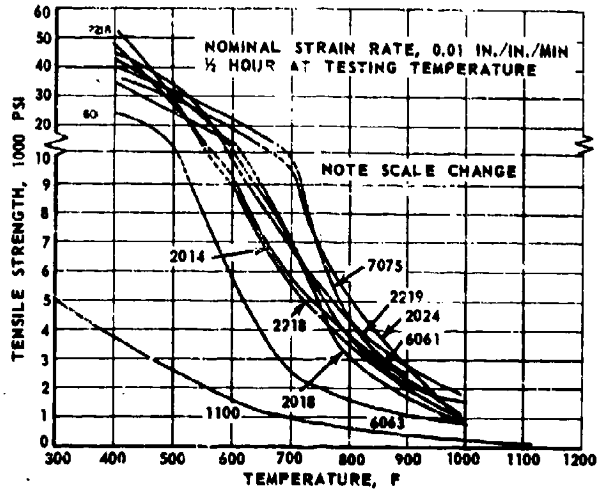


FIGURE 7-5. ELEVATED-TEMPERATURE TENSILE STRENGTHS FOR ALUMINUM AND SEVERAL ALUMINUM ALLOYS(6)

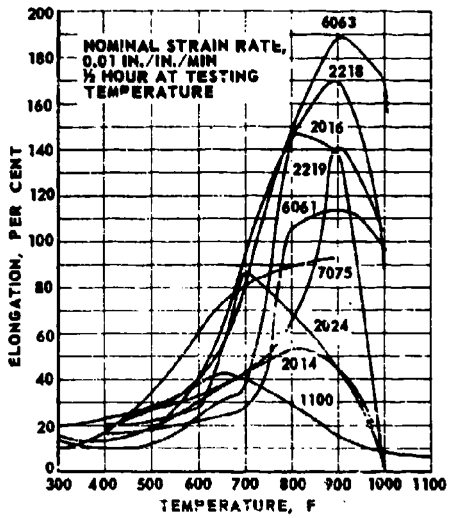


FIGURE 7-6. ELEVATED-TEMPERATURE ELONGATIONS FOR ALUMINUM AND SEVERAL ALUMINUM ALLOYS(6)

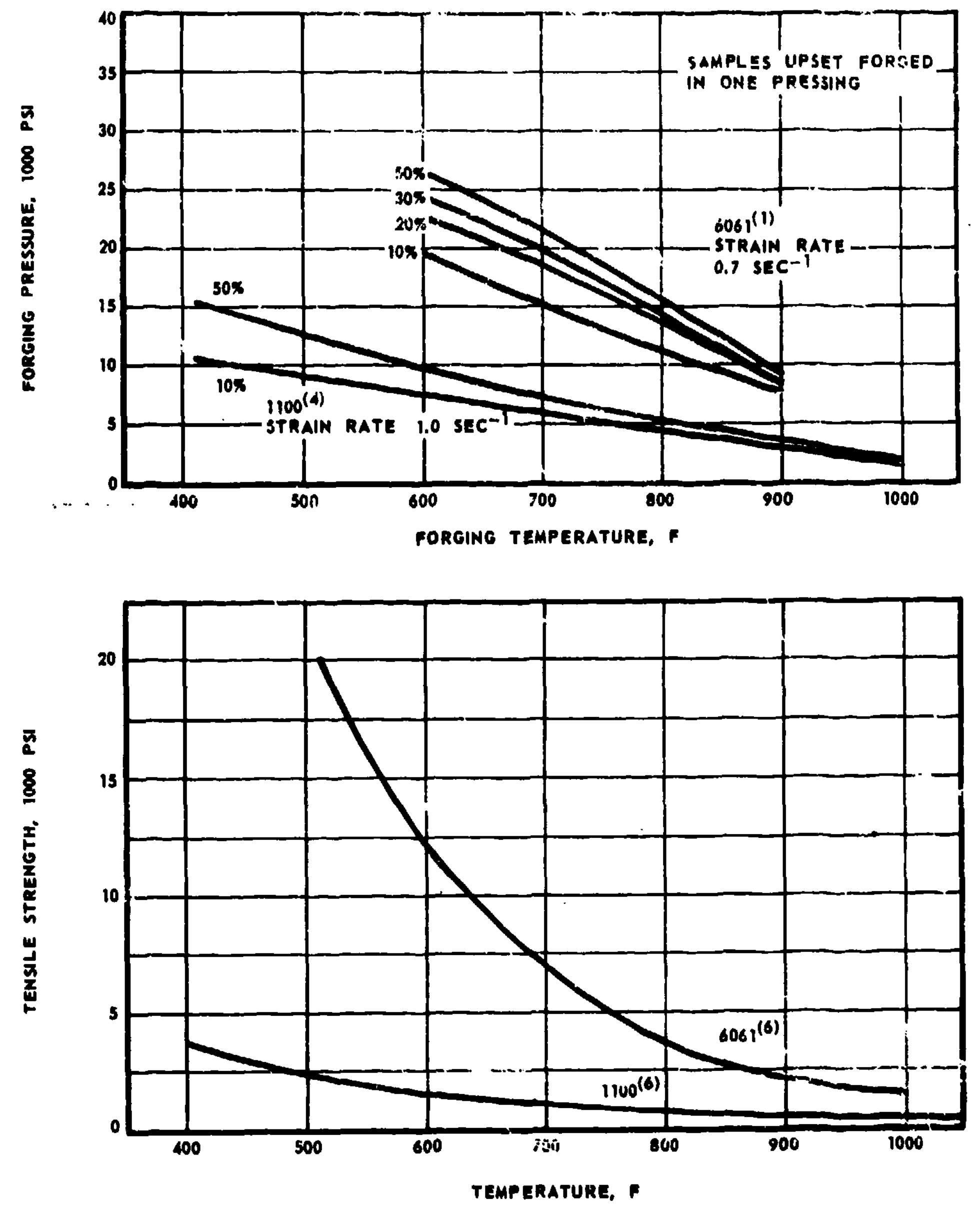


FIGURE 7-7. FORGING PRESSURE AND TENSILE STRENGTH AS FUNCTIONS OF TEMPERATURE FOR 1100 AND 6061 ALUMINUM[1,4,6]

and, to a certa[illegible] precipitation hardening that occurs at the working temperatures. When th[illegible] [illegible]tion heat treated, the precipitating phases are adequately dissolved and [illegible] [illegible]ed to a minimum. Maximum strength then is obtained by aging. Alu[illegible] [illegible]rely used in the as-forged condition.

The forgi[illegible] [illegible]wo conditions which influence the heat-treated properties of al[illegible] [illegible]ow and final grain size. Wrought aluminum-alloy bars gene[illegible] [illegible] ductility, the maximum in the longitudinal direction of the b[illegible] [illegible]n the transverse direction. The transverse ductility can be in[illegible] [illegible]the bar is deformed to produce flow in the transverse or lateral di[illegible] [illegible] depicts the typical tensile-elongation values of alumin[illegible] [illegible]after forging to impart increasing amounts of lateral flow. Longitudinal [illegible] [illegible]ains essentially constant while the long-transverse elongation increases [illegible] [illegible]easing reduction. The short-transverse elongation either remains essentially constant or it will decrease slightly if the reduction is exceedingly high. Lowest short-transverse elongation thus occurs adjacent to the parting line (see section on Grain Flow).

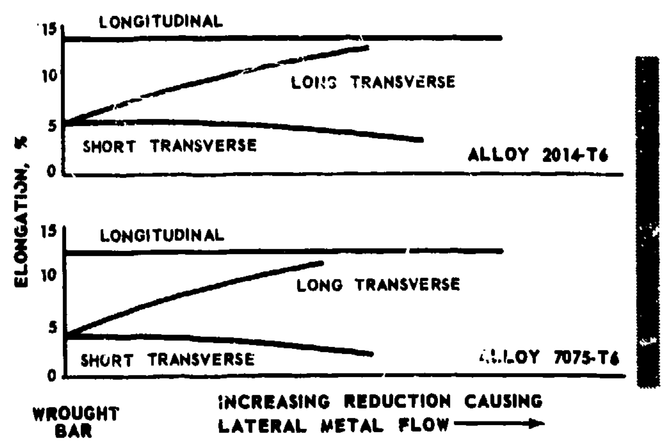


FIGURE 7-8.

INFLUENCE OF FORGING REDUCTION ON THE TENSILE ELONGATION OF ALUMINUM ALLOYS

COMMENTARY ON FORGING PRACTICES

Out of the family of structural metals and alloys, the aluminum alloys are most readily forged to precise, intricate shapes. The most significant reasons for this are: (1) the alloys are ductile, (2) they can be forged in dies heated to essentially the same temperatures as the workpiece, (3) they do not develop scale during heating, and (4) they require low forging pressures.

The major factors influencing the forgeability of aluminum alloys are the solidus temperature and deformation rate. Most of the alloys are forged at about 100 F below the solidus temperature. The risk of incipient melting exists when conditions of forging promote significant temperature increases. Such increases are caused by either too rapid or too large forging reductions. If incipient melting occurs, the forgings will sometimes rupture and will be permanently damaged. To minimize this risk, forging temperatures are often adjusted downward for increasing amounts and/or increasing rates of reduction. Examples of how forging temperatures should be adjusted for different conditions are given in Table 7-3. The 7075 and 7079 alloys exhibit poorer forgeability in hammers than in presses and thus are usually press forged. Cross

forging (a series of upset and drawing-out operations conducted between flat dies) usually results in significant temperature increases; hence, the temperatures are lower than for ordinary hammer and press forging.

TABLE 7-3. ADJUSTMENT OF FORGING TEMPERATURES TO AVOID CRACKING OF ALUMINUM ALLOYS IN VARIOUS FORGING OPERATIONS[1]

Alloy	Solidus Temperature, F	Forging Temperature, F, for Given Types of Forging Operations: Die Forging in Hydraulic Presses	Die Forging in Drop Hammers	Upset Forging in Drop Hammers	Cross Forging in Presses	High-Energy-Rate Machine
2014	950	850	750-830	-- 750	710	650
2025	970	880	800-850	--	750	--
6061	1080	880	850	--	750	--
7075	890	750	700-725	700	650-700	650
7079	900	750	700-725	--	650-700	--

Aluminum alloys can be forged in any of the forging-equipment types used for other metals. Single-stroke presses are often preferred for forging shapes requiring large reductions, since lubrication can be applied easily and the pressing speeds are low enough to avoid harmful temperature increases. Since aluminum does not form a scale when heated, direct metal-to-die contact causes seizing and galling. Hence, it is important to have good, continuous lubrication.

Lubrication. Lubricants used for aluminum-alloy forgings vary from kerosene, to oil-graphite suspensions, to complex, proprietary commercial compounds. Forging companies frequently mix many of these compounds with one or more oils and some companies permit individual press operators to slightly modify the mixtures for ease of application. Hence, it is difficult to apply a precise rating to the many possible lubricants. In general, the graphite suspensions applied by spraying are preferred for press forging, while water-soluble soaps applied by swabs are frequently used for hammer forging. Another technique consists of dipping aluminum forging blanks in caustic (10% NaOH) to produce a porous conversion coating on the surface. The forgings are then dipped in a colloidal graphite dispersion and dried prior to forging. This technique eliminates many of the seizing and galling problems but adds the problem of corrosion if the forgings are not cleaned soon after forging.

Residual Stress. A significant problem with heat-treatable aluminum forgings is associated with the residual stresses imposed by quenching, and sometimes by straightening just after quenching. The aging temperatures are usually too low to adequately remove the residual stresses which, in some cases, exceed 30,000 psi. Residual stresses often cause warpage during machining and promote stress corrosion. Partially successful approaches to these problems include: (1) quenching into boiling water to minimize residual stresses, (2) cold coining or peening to lower the residual stresses, and (3) reverse quenching or cooling solution-heat-treated forgings slowly to extremely low temperatures after quenching, then reheating them rapidly by steam, hot water, hot dies, etc. The theory behind these methods is that residual stresses can be minimized by imposing stresses essentially equal but opposite to the original quenching stresses.

Blisters. Aluminum alloys are, at times, subject to the formation of blisters or voids near the forging surface. These defects are usually observed after solution heat treatment. There are at least three different types of blisters found on forged aluminum. When the blisters are opened with a pointed tool, the inside surfaces appear either gray, dark brown or black, or white. These colors generally identify blisters caused, respectively, by (1) incipient melting, (2) improper lubrication, or (3) dissolved gasses. The first type are observed where solution heat treating temperatures are too high. The second type of blisters form when inadequate lubrication causes the surface metal to seize and shear away from the subsurface metal. Lubricants trapped below the surface volatilize during heat treatment, causing the blisters. These blisters are most frequently found on edges of blades and portions of other forgings characterized by rapid metal flow.

The third type of blister is not fully understood. The most popular theory holds that atomic hydrogen, accumulated during melting, fabricating, and heat treating, concentrates in lattice imperfections, and then changes to the less-soluble molecular form. Further it is believed that the blisters form when the trapped gas expands during the solution-annealing cycle. Whatever the true origin for the gas-type blisters, they can be prevented by applying commercial, water-soluble corrosion inhibitors to the forgings before heat treatment.

COMMENTARY ON FORGING DESIGN

In principle, aluminum alloys can be forged to any shape consistent with limits set by die design. In the last decade, for example, forging companies developed a no-draft precision forging that requires little or no preassembly machining. The tooling costs for such forgings are considerably higher than those for more conventional designs. For this reason, precision forgings are specified only where the tooling costs can be amortized over large production quantities. An often quoted guide is that a production order should be in the neighborhood of 500 forgings for precision forging to be economical. This figure will vary depending on the balance between final machining-cost reductions and tooling-cost increases. Most forging companies will provide estimates for both conventional and precision designs so that a direct cost comparison can be made. A typical no-draft precision forging of aluminum is shown in Figure 7-9.

Forging Size and Shape. The Air Force's heavy-press program has made possible the closed-die forging of aluminum-alloy parts up to 20 feet in length. Vertical ribs on the order of 1/8 inch thick with zero draft have been forged successfully on aluminum parts weighing over 50 pounds. These same parts have thin webs ranging from 3/16 to 1/4 inch thick. Maximum limits on forging size depend a great deal on the precision requirements and final techniques, but, as a general rule, the maximum forging size (plan area) is in the neighborhood of 3500 square inches for aluminum die forgings.

Excess Stock Allowances. Since aluminum alloys are not sensitive to scale formation or contamination, excess stock allowances are necessary only to obtain dimensional accuracy and provide for localized deviations. Precision forgings are often forged to final size and require no excess stock allowances.

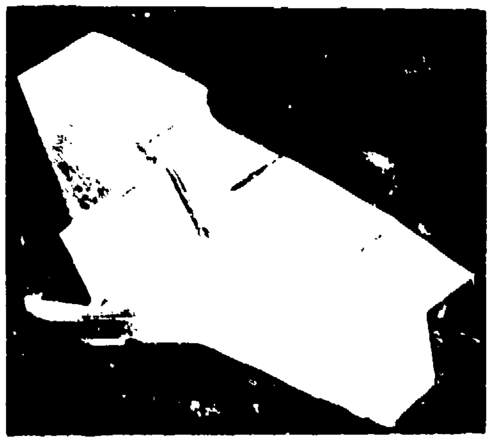


FIGURE 7-9. A TYPICAL NO-DRAFT PRECISION FORGING OF ALUMINUM ALLOY

Courtesy of Wyman-Gordon Company

Metallurgical Factors in Design. Since some of the alloys (notably 6061 and 2014) are subject to excessive grain growth, the problems of grain-size control require carefully balanced reductions, forging temperatures, and die temperatures. Frequently, close-tolerance forgings receive only small reductions during the final stages of forging at comparatively low temperatures and are likely to exhibit abnormal grain growth upon subsequent solution heat treatment. This is rarely a problem with forgings more conventional in design.

REFERENCES

(1) Henning, H. J., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", Battelle Memorial Institute, Technical Documentary Report No. ML-TDR-64-95, Contract No. AF 33(600)-42963 (March, 1964).

(2) Designing for Alcoa Forging, Aluminum Company of America, Copyright 1950.

(3) Metals Handbook, 8th Edition, Vol 1, Properties and Selection of Materials, American Society for Metals, Cleveland, Ohio, Copyright 1961.

(4) Alder, J. F., and Phillips, U. A., "The Effects of Strain Rate and Temperature on the Resistance of Aluminum, Copper, and Steel to Compression", J. Inst. of Metals, 83, 80-86 (1954-1955).

(5) Kent, W. L., "The Behavior of Metals and Alloys During Hot Forging", Trans. AIME, 34 (1928).

(6) Voorhees, H. R., and Freeman, J. W., "Report on the Elevated-Temperature Properties of Aluminum and Magnesium Alloys", ASTM Publication 291, Philadelphia, Pennsylvania (October, 1961).

CHAPTER 8

MAGNESIUM FORGING ALLOYS

Crystalline Structure:	Close-packed hexagonal
Phase Changes:	None
Number of Phases:	Generally one phase at forging temperatures
Liquidus Temperature Range:	1130 to 1200 F
Solidus Temperature Range:	914 to 1198 F
Reactions When Heated in Air:	Slow oxidation; pyrophoric when heated above 900 F

ALLOYS AND FORMS AVAILABLE

Table 8-1 gives the composition of magnesium alloys frequently used for die forgings. Included are data useful for predicting forging behaviors.

TABLE 8-1. COMPOSITIONS AND FORGING CHARACTERISTICS OF SEVERAL MAGNESIUM ALLOYS(1, 2, 3, 4, 5)

Alloy Designation	AZ31	AZ61	AZ80	ZK60	HM21	HM31	HK31
Forging Specifications	52S	AMS 4358 57S	AMS 4360 AMS 4362 58-S	AMS 4362	--	AMS 4388	--
Nominal Composition, %	3.0Al 1.0Zn	5.5Al 1.0Zn	8.5Al 0.5Zn	5.5Zr 0.5Zn	2.0Th 0.5Mn	3.0Th 1.2Mn	3.0Th 0.1Zn
Solidus, F	1120	977	915	970(a)	1120	1120	1100
Recrystallization Temp, F	400	500	600	600	--	--	--
Forging-Temp Range, F	600-800	600-800	550-775	600-750	750-1050	750-1050	600-1050
Normal Forging Temp, F(b)							
Presses	750	750	750	675	950	950	900
Hammers	--	--	--	--	--	900	--
Relative Forgeability							
Presses	Good	Good	Good	Excellent	Good	Good	Good
Hammers	Poor	Poor	Poor	Poor	Fair	Fair	Fair

(a) As-cast ZK60 may, at times, contain traces of eutectic composition that liquates at 645 F causing poor forgeability at higher temperatures.

(b) For alloys sensitive to rapid grain growth at forging temperatures, it is common practice to conduct each forging step at successively lower temperatures.

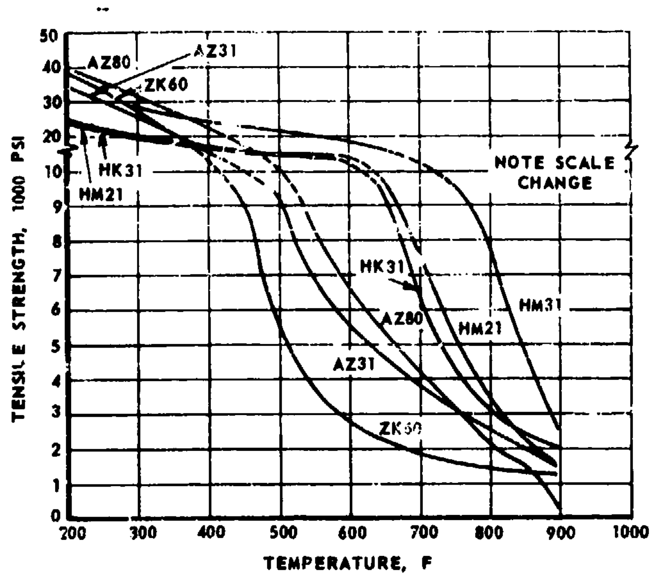


FIGURE 8-1. ELEVATED-TEMPERATURE TENSILE STRENGTHS FOR SEVERAL MAGNESIUM ALLOYS(3,5)

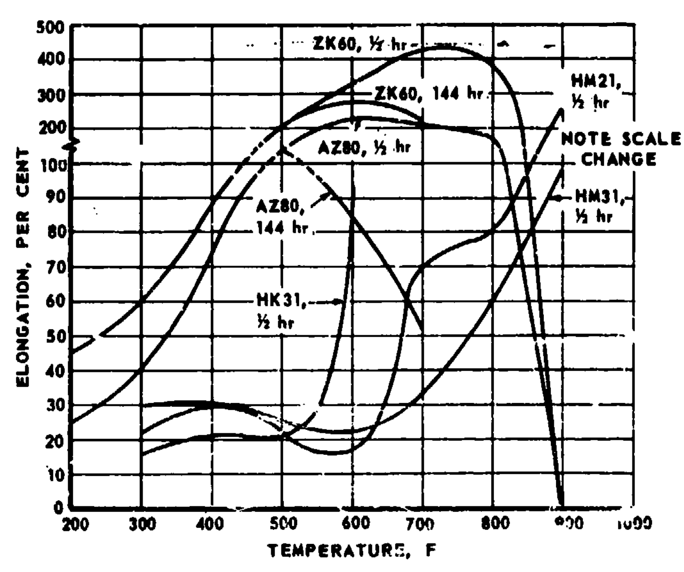


FIGURE 8-2. ELEVATED-TEMPERATURE ELONGATIONS FOR SEVERAL MAGNESIUM ALLOYS(3,5)

Times indicated are total time at temperature before testing.

Ingots are generally converted to large billets by extrusion. Smaller bars are hot rolled from the extruded billets. Continuously cast ingots are sometimes forged directly.

FORGING BEHAVIOR

Figures 8-1 and 8-2 give tensile strength and elongation values, respectively, for several magnesium forging alloys at elevated temperatures. The alloys containing thorium retain strengths at higher temperatures than do the alloys containing aluminum, zinc, and zirconium and thus require higher forging pressures. The significant changes in elongation with temperature are useful for indicating forging-temperature ranges. The AZ80 alloy, for example, should be forged above 500 F where elongation is high. The HM21 alloy should be forged above 800 F where a substantial increase in elongation occurs. These temperatures correspond approximately with minimum temperatures established in forging practice.[10]

Elongation data for the AZ80 alloy show how extended time at temperatures above 500 F decreases ductility because of grain growth. ZK60, on the other hand, is far less sensitive to grain growth. In practice, the greatest problem of grain growth during heating for forging is found with the AZ80 alloy.

Forging pressure data, determined by upsetting billets between flat dies, are presented in Figure 8-3 for three magnesium alloys at normal temperatures. At normal pressing speeds, the forging pressure increases at first and then drops slightly, probably because the temperature increases during forging. The curves for AZ31 show that the effect of strain rate on forging pressure varies with reduction. It is interesting that the forging pressures for each of these alloys is about 3 to 4 times greater than the values for tensile strength given in Figure 8-1.

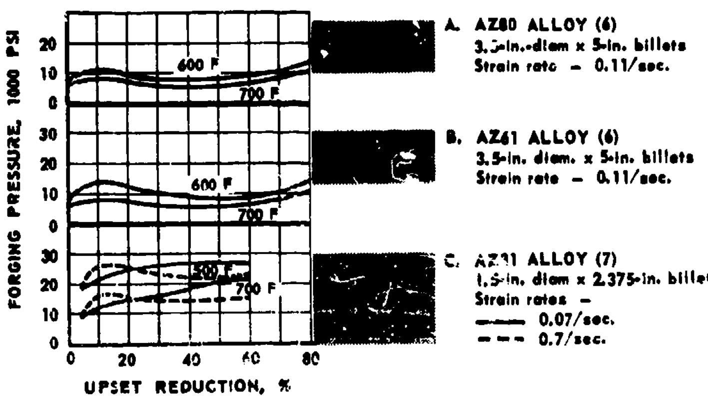


FIGURE 8-3. FORGING-PRESSURE CURVES FOR THREE MAGNESIUM ALLOYS UPSET BETWEEN FLAT DIES[6,7]

Forging pressures for die forgings containing thin webs (0.20-0.22-inch thick) are on the order of 60,000 psi for parts made from the HM21 alloy at 850 F[2]. This represents about 6 times the flow stress of the alloy at that temperature. In other words, forging pressures for die forgings may in many cases considerably exceed the values indicated by the tensile strength given in Figure 8-1.

Forgeability of magnesium alloys is influenced by three important factors: solidus temperature, deformation rate, and grain size. Like aluminum alloys, magnesium alloys are forged at temperatures that are often within 100 F of the solidus temperature. The high-zinc magnesium alloy (ZK60), sometimes will contain small amounts of the low-melting eutectic that forms during ingot solidification. This eutectic melts upon heating to temperatures just over 600 F and can cause severe rupturing when the alloy is forged above this temperature. To minimize this problem, mill suppliers generally homogenize cast ingots at elevated temperatures for extended periods to redissolve the eutectic and to restore the higher solidus temperature. The commercial alloys containing aluminum, zirconium, and thorium do not form low-melting eutectics and, therefore, are not subject to hot shortness at such low temperatures.

Wrought magnesium alloys exhibit excellent forgeability in slow-moving hydraulic presses but often crack when forged in drop hammers. Even in presses, the faster-moving flash metal is sensitive to cracking. Billets containing coarse grains are particularly sensitive to cracking during rapid deformation. For this reason, coarse-grained ingots are generally extruded before forging.

INFLUENCE OF FORGING VARIABLES ON MECHANICAL PROPERTIES

The room-temperature mechanical properties of magnesium alloy forgings, especially ductility, are strongly dependent on forging procedures. In general, both longitudinal and transverse ductility improve with decreasing grain size and with increasing amounts of work. The following data show how longitudinal and transverse elongations are affected by grain size in typical AZ80 alloy forgings:[7]

Grain Size	Elongation, %	
	Longitudinal	Transverse
Coarse ASTM 0-1	3-4	1-2
Fine ASTM 5-7	9-12	5-7

While the basic strength properties of magnesium alloys are determined by alloy composition, forging plays an important role in establishing property uniformity and maximum ductility.

An important objective during forging of magnesium alloys is to refine the grain size. This is done through careful control of both forging temperatures and forging reductions. It is common forging practice to forge magnesium alloys at successively lower temperatures for each forging operation. This procedure retards grain growth during forging and promotes maximum grain refinement. Rosenkranz[8] studied the influence of forging temperatures and forging reductions on the grain size of MgAl9 (German equivalent to AZ80) after single reductions and after multiple-step reductions. As expected, grain size decreased with increasing reductions and with

decreasing forging temperatures when forged in a single pressing. Other specimens were given three and four successive reductions under following conditions:

(1) At constant temperatures

(2) At progressively lower temperatures (50 F per step from 700 F)

(3) At progressively higher temperatures (50 F per step from 550 F).

The finest grains (2-5 mm dia.) were obtained when forging was done at progressively lower temperatures except when reductions were less than 10 per cent. For reductions less than 20 per cent the finest grains (2-4 mm dia.) were obtained by forging in multiple steps at 550 F. The grain size of specimens given multiple reductions, at the same temperature, increased with forging temperature as follows:

Forging Temperature, F	Grain Diameter After 2-3 Successive Reductions, mm
550	2-4
590	3-6
640	6-9
700	14-19

Forging with progressively increasing temperatures led to variable grain sizes ranging from 12 to 24 mm in diameter. Thus, large differences in grain size can result from comparatively small changes in forging temperature and reduction schedules. Hence, the control of these factors is very important to achieving satisfactory mechanical properties. The same principles apply to many other alloys.

It is also important to provide as much flow in the transverse direction as possible during forging. This is necessary because wrought magnesium-alloy bars exhibit highly directional ductility. By providing transverse metal flow, the transverse ductility is improved and the longitudinal ductility is kept high. The following transverse-elongation values are typical of the AZ80 alloy:

Condition	Elongation, %	
	Longitudinal	Transverse
Alloy bar	10	3
Forging receiving small transverse metal flow (about 15 per cent reduction)	10-11	4-5
Forging receiving severe transverse metal flow (about 60 per cent reduction)	11-13	6-8

Mechanical properties of three different alloys determined on parts forged to the same configuration are presented in Table 8-2. These data demonstrate how variations in mechanical properties are influenced by grain-flow orientation and the relative amount of deformation imparted during forging. These variations are quite similar

for each of the alloys. In all cases, the sections combining the greatest reductions with longitudinal grain flow exhibited the highest strength values.[9] Similar conclusions were drawn from forgings produced from the HM21 and HK31 alloys.[10]

TABLE 8-2. TYPICAL MECHANICAL PROPERTIES OF COMMERCIAL MAGNESIUM-ALLOY AIRCRAFT FORGINGS

Alloy	Grain-Flow Orientation[a]	Test Location	Relative Amount of Deformation[b]	Yield Strength (0.2% Offset), 1000 psi	Ultimate Tensile Strength, 1000 psi	Elongation, per cent
ZK21	Long.	Rim Tang.	Moderate	25.6	38.7	25.5
(as forged)	Long.	Radial	Large	25.9	38.9	17.8
	Long.	Radial	Large	27.6	40.0	20.2
	Long.	Rim Vertical	Moderate	19.0	32.1	20.2
	Long.	Center Vertical	Light	14.6	33.9	20.4
	Trans.	Center Tang.	Moderate	22.7	37.3	20.0
AZ61	Long.	Rim Tang.	Moderate	26.5	41.2	20.0
(as forged)	Long.	Radial	Large	25.4	40.4	16.2
	Long.	Radial	Large	24.6	40.3	17.4
	Long.	Rim Vertical	Moderate	17.1	37.5	14.0
	Long.	Center Vertical	Light	13.8	38.2	18.6
	Trans.	Center Tang.	Moderate	23.0	39.4	17.7
AZ31	Long.	Rim Tang.	Moderate	25.9	39.3	18.0
(as forged)	Long.	Radial	Large	21.1	37.8	18.0
	Long.	Radial	Large	24.5	38.2	16.6
	Long.	Rim Vertical	Moderate	17.1	36.2	11.6
	Long.	Center Vertical	Light	14.8	35.9	18.0
	Trans.	Center Tang.	Moderate	23.7	38.1	18.0

Data courtesy of Wyman-Gordon Company.

(a) Long. and Trans., respectively, denote longitudinal or transverse grain flow in test bars.

(b) In this forging, metal flows in three directions, hence, values for deformation are comparative only.

COMMENTARY ON FORGING PRACTICES

Procedures for forging magnesium alloys are similar in many respects to those used for aluminum. Presses are favored for forging, and die-temperature control is most important. If die temperatures are much below forging temperatures, surface cracking is a potential problem. When die temperatures exceed forging temperatures the metal flows past impressions and promotes underfilling.

Alloys requiring forging temperatures above 900 F must be heated in inert or reducing atmospheres to prevent burning. Sulfur dioxide gas is frequently used for this purpose.

Lubrication. The most popular die lubricants for magnesium are lamp black, applied directly by an oil or kerosene torch, and colloidal graphite (water base). Light oil suspensions of graphite are frequently applied directly to critical projections on forgings to reduce friction in the die cavity. Care is taken to avoid application of such lubricants to the flash region where friction is necessary. Fluid lubricants, e.g., greases, oils, etc., are seldom used because they promote rupturing.

Die Heating. As a general rule, die temperatures are maintained at temperatures corresponding approximately with forging temperatures. To maintain close control of die temperatures, thermostatically controlled gas-burner rings are generally used. As a safety measure, shields are often used to keep the flame away from the impression. Fire is a particular hazard when forging the thorium-containing alloys at temperatures near 900 F.

Grain-Size Control. Alloys subject to rapid grain growth at forging temperatures (AZ31, AZ61, AZ80) are generally forged at successively lower temperatures for each operation. Common practice is to reduce the temperature about 25 to 30 F after each step. It is sometimes a preferred practice for parts containing regions that receive only small reductions to conduct all forging steps at the lowest practical temperatures (e.g., 625 F for AZ61 and 600 F for AZ80). Grain growth in such alloys as ZK60, HK31, and HM21 is slow at forging temperatures, and these alloys may be forged in multiple operations at constant temperatures without risking extensive grain growth.

Trimming. Because magnesium alloys are comparatively brittle during cold trimming, flash is generally removed from forgings by band sawing. Magnesium forgings may be warm trimmed at temperatures corresponding to minimum forging temperatures. This may lead to other problems of bending and warping. For these reasons warm trimming is generally confined to forgings with comparatively heavy sections in the flash-line regions.

COMMENTARY ON FORGING DESIGN

The design of magnesium-alloy forgings is similar in many respects to that for aluminum. Magnesium does not flow as readily into deep vertical cavities. Greater fillet radii are generally provided and flash openings are usually narrower. For these reasons, the more complex magnesium forgings often require an additional preliminary die. For example, if two dies are needed for a typical aluminum structural forging, a comparable magnesium forging will require three dies for successful forging.

In principle, the limitations on sizes of magnesium-alloy forgings are essentially the same as those for aluminum alloys. However, because of little demand for extremely large forgings there has not been much incentive to increase the size potential beyond about 2000 sq. in. of plan area. The largest magnesium forging mentioned in a recent survey was a 229-pound hemispherical forging used as a canister for the ECHO satellite program. Thin ribs and webs, on the order of 1/8 inch thick, have been forged from magnesium alloys. Tolerances of ±0.010 inch have been held successfully on small parts.

A typical close-tolerance, magnesium-alloy forging, illustrating typical size and shape capability, is shown in Figure 8-4. This part was made by conventional techniques in a hydraulic press. Figure 8-5 shows a no-draft fan-blade forging. This part was extrusion forged in a hydraulic press equipped with concentric rams. The operations combined both hot-die and split-die forging techniques.

FIGURE 8-4. TYPICAL CLOSE-TOLERANCE MAGNESIUM-ALLOY DIE FORGING

Courtesy of Wyman-Gordon Company

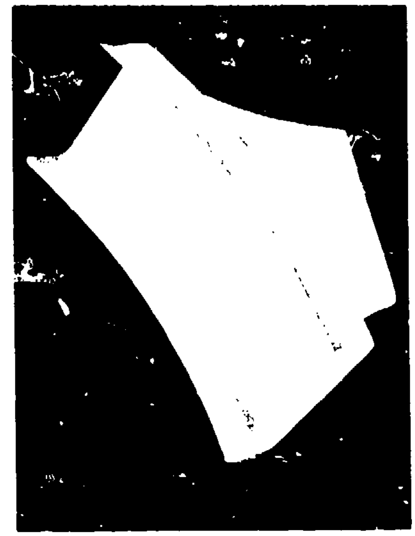


FIGURE 8-5. NO-DRAFT FAN BLADE FORGING PRODUCED BY EXTRUSION FORGING IN PRESS EQUIPPED WITH CONCENTRIC RAMS

Courtesy of Wyman-Gordon Company

REFERENCES

(1) Metals Handbook, 8th Edition, Vol. 1, "Properties and Selection of Materials", American Society for Metals, Cleveland, Ohio, Copyright 1961.

(2) Lockheed Aircraft Corporation, "Forging Magnesium Alloys", Preliminary Information on Air Force Contract AF 33(600)-36577.

(3) Voorhees, H. R., and Freeman, J. W., "Report on the Elevated-Temperature Properties of Aluminum and Magnesium Alloys", ASTM Publication 291, Philadelphia, Pennsylvania (October, 1961).

(4) Dow Metal Handbook, Dow Chemical Company, Midland, Michigan, Copyright 1944.

(5) Craighead, C. M., Grube, K. R., Eastwood, L. W., and Lorig, C. H., "The Effects of Temperature on the Mechanical Properties of Magnesium Alloys", A Technical Survey by Battelle Memorial Institute for Rand Corporation (April, 1949).

(6) Dietrich, R. L., and Ansel, G., "Calculation of Press Forging Pressures and Applications to Magnesium Forgings", Trans. ASM, 38, 709-727 (1947).

(7) Henning, H. J., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", Technical Documentary Report No. ML-TDR-64-95, Contract No. AF 33(600)-42963, Battelle Memorial Institute (March, 1964).

(8) Rosenkranz, W., "The Forging and Die Forging of Magnesium Alloys of the Mg-Al-Zn Type", Zeitschrift für Metallkunde, 47, 107-117 (1956).

(9) Jablonski, S. J., "Magnesium Forging Developments Provide Greater Economics and Wider Application Potential", Paper presented at the 17th Annual Convention of the Magnesium Association, New York (October 16-18, 1961).

(10) Pashak, J. F., "Forging Characteristics and Properties of HM21XA and EK31XA Magnesium Alloy Production Forgings", Dow Chemical Company, WADC TR 58-218, AD 204797 (November, 1958).

CHAPTER 9

CARBON AND LOW-ALLOY STEELS

Crystalline Structure:	Austenite – face-centered cubic Ferrite – body-centered cubic
Phase Changes:	Austenite tranforms to ferrite plus carbide over a range of temperatures below 1500 F
Number of Phases:	One phase (austenite) at forging temperatures
Liquidus Temperature Range:	2650 to 2750 F
Solidus Temperature Range:	2450 to 2700 F
Reactions When Heated in Air:	Oxidation, rapid above 1400 F; decarburization

ALLOYS AND FORMS AVAILABLE

There are hundreds of steels that range in composition from about 0.10 to 1.0 per cent carbon and from trace amounts to about 5 per cent each of several metallic alloying elements (e.g., Cr, Mo, V, Ni, etc.). The carbon content controls the hardness and strength levels obtainable, while the alloying serves the purposes of increasing hardenability and hot strength, improving resistance to thermal and mechanical shock, providing grain refinement, and improving machinability. The steels in this group range from standard carbon and low-alloy steels to specially alloyed, superstrength steels like 300M, D6AC, and H-11. At their respective forging temperatures, all of these alloys exhibit the same basic forging behavior.

Raw material used for forging is generally bar hot rolled from ingots melted in open-hearth, electric-arc, or vacuum furnaces. Thus, billet material has usually received considerable reduction before the die-forging operations. For certain grades, vacuum melting imparts better forgeability than does conventional arc melting. However, the major purpose of vacuum melting is the improvement of mechanical properties, not forging behavior.

FORGING BEHAVIOR

Forgeability

The extent to which carbon and low-alloy steels can be forged into intricate shapes is seldom limited by forgeability problems, except when the grades contain sulfides, bismuth, or other intentional additions for other purposes (e.g., free machining). Probably the most important factor limiting the section thickness, shape

complexity, and forging size is the cooling that occurs when the heated work-piece comes in contact with colder dies. This is essentially the reason that more intricate shapes can be forged in hammers than in presses, under otherwise similar conditions of stock and die temperatures.

Ihrig[1] and Clark and Russ[2] studied the relative forgeability of numerous steels using hot-twist tests. Twist data obtained on carbon and low-alloy steels (Figure 9-1) generally show that forgeability increases with increasing temperature until the melting point of the steels is approached, and that maximum forging temperature decreases with increasing carbon content. Figure 9-1 shows a comparison of twist data for plain carbon steel and several low-alloy steels of the same carbon content with twist data for a leaded steel. The deleterious effect of lead additions on twist ductility and, thus, forgeability is quite apparent.

The influence of several individual alloying elements on the forgeability of steels is shown by the hot-twist data in Figures 9-2 through 9-5. In addition to lead, tin and sulfur in small amounts have adverse effects on forgeability. Sulfur is especially harmful in steels lean in manganese. Chromium and molybdenum additions also reduce forgeability. Small additions of nickel and manganese improve forgeability, while the reverse is true for additions beyond certain limits.

In general, the forgeability of steels improves with increasing deformation rate. Anderson[3] has shown this effect in twist tests where the number of twists to failure for low-carbon steel increased with increasing twisting rate (Figure 9-6). It is believed that the improvement in workability is due primarily to heating effects that occur with the increasing deformation rates.

Flow Stress and Forging Pressure

Comparative flow stresses and forging pressures for many carbon and low-alloy steels are obtainable from hot-twist tests and elevated-temperature tensile and compression tests. Figure 9-7 gives torque versus temperature curves for several carbon and low-alloy steels which demonstrate that the relative forging-pressure requirements for the various alloys in this group do not differ widely at normal forging temperatures around 2200 F. A torque curve for Type 304 stainless steel is included to show how higher alloy contents increase flow strengths.

Actual forging-pressure measurements on AISI 1020 and 4340 steels and a Type A6 medium-alloy tool steel for upsetting reductions of 10 and 50 per cent over a range of temperatures are given in Figure 9-8. These data show that the pressures required for forging 1020 and 4340 under the same set of conditions vary only slightly, as predicted from the torque curves. The more highly alloyed A6 tool steel, on the other hand, requires considerably greater forging pressures. In addition, this alloy exhibits a more significant increase in forging pressure with increasing reduction.

Forging pressures for a given steel increase with increasing deformation rates. Upset forging studies on low-carbon steel by Alder and Phillips[5] showed that the influence of strain rate was more pronounced at higher forging temperatures. These effects are shown graphically in Figure 9-9 which gives the stress-strain curves for a low-carbon steel upset at various temperatures over a 6-fold range of strain rates.

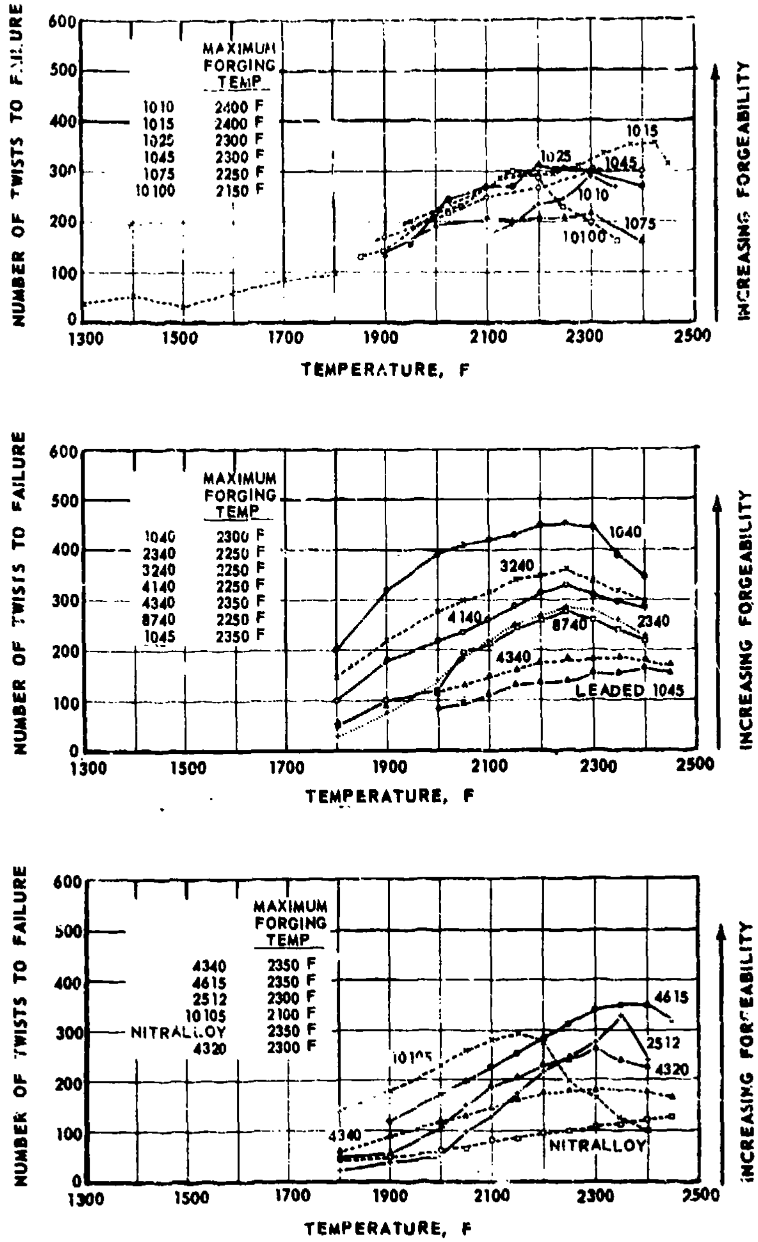


FIGURE 9-1. RELATIVE FORGEABILITIES OF STEELS AS DEMONSTRATED BY HOT-TWIST TESTS

After Ihrig[1] and Clark and Russ[2]

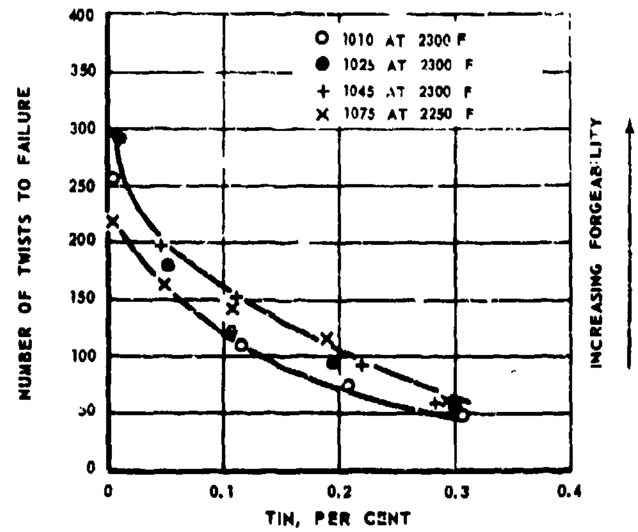


FIGURE 9-2. INFLUENCE OF TIN ON FORGEABILITY OF CARBON STEELS

After Ihrig[1]

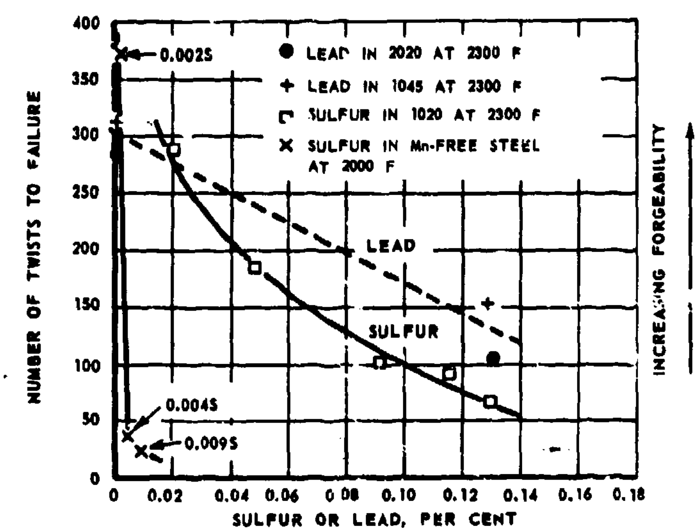


FIGURE 9-3. INFLUENCES OF LEAD AND SULFUR ON FORGEABILITY OF CARBON STEELS

After Ihrig[1] and Anderson, et al.[3]

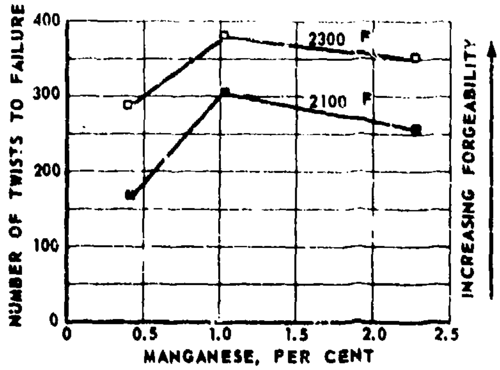


FIGURE 9.4.

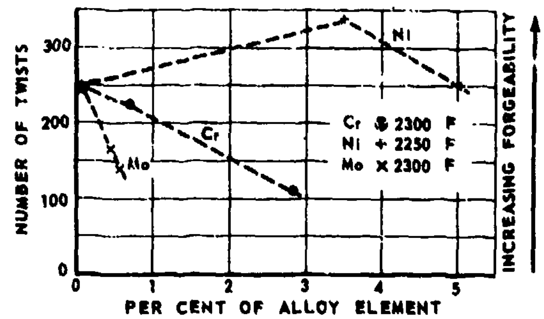
INFLUENCE OF MANGANESE ON FORGEABILITY OF 1020 STEEL CONTAINING NOMINALLY 0.025 SULFUR

After Ihrig[1]

FIGURE 9.5

INFLUENCE OF NICKEL, CHROMIUM, AND MOLYBDENUM ON FORGEABILITY OF STEELS CONTAINING NOMINALLY 0.15 CARBON

After Ihrig[1]



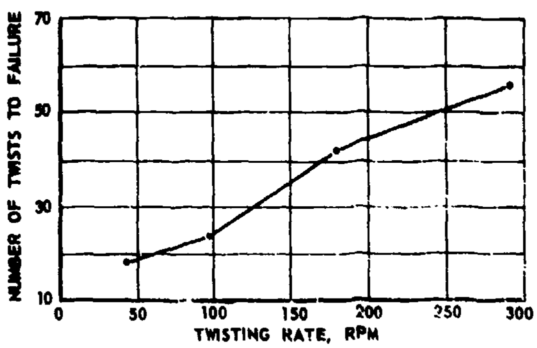


FIGURE 9.6.

INFLUENCE OF DEFORMATION RATE ON HOT-TWIST CHARACTERISTICS OF LOW-CARBON STEELS AT 2000 F

After Anderson, et al.[3]

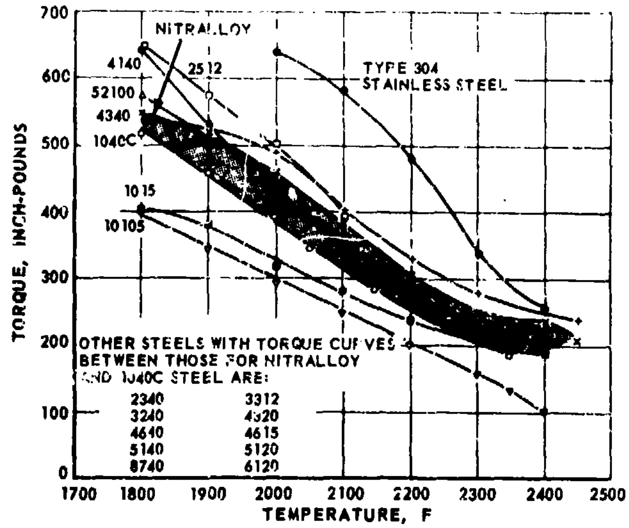


FIGURE 9-7. TORQUE CURVES FOR SEVERAL CARBON AND LOW-ALLOY STEELS SHOWING HOW RELATIVE FORGING-PRESSURE REQUIREMENTS VARY WITH TEMPERATURE

After Clark and Russ[2]

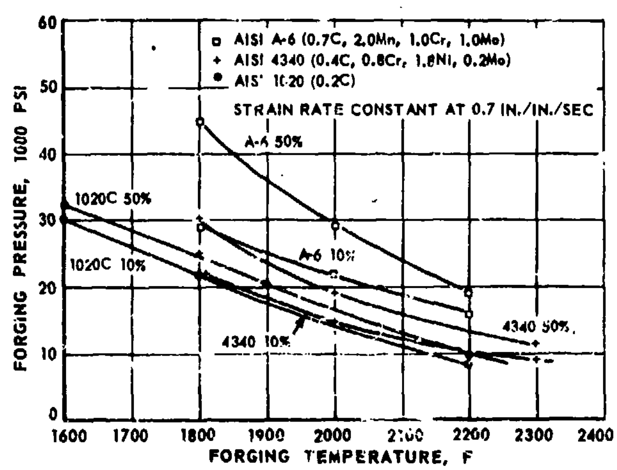


FIGURE 9-8. FORGING PRESSURE FOR THREE STEELS UPSET AT VARIOUS TEMPERATURES AND REDUCTIONS[4]

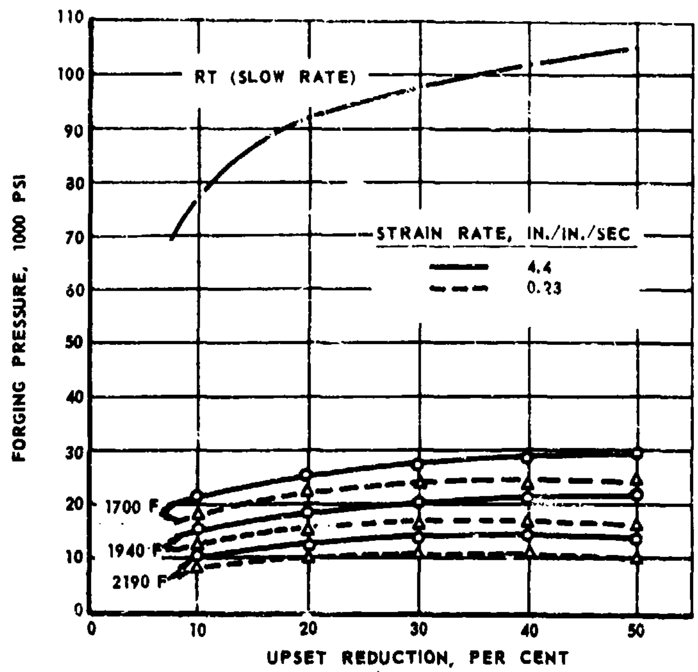


FIGURE 9-9. FORGING PRESSURE FOR MILD STEEL UPSET AT VARIOUS TEMPERATURES AND TWO STRAIN RATES

After Alder and Phillips[5]

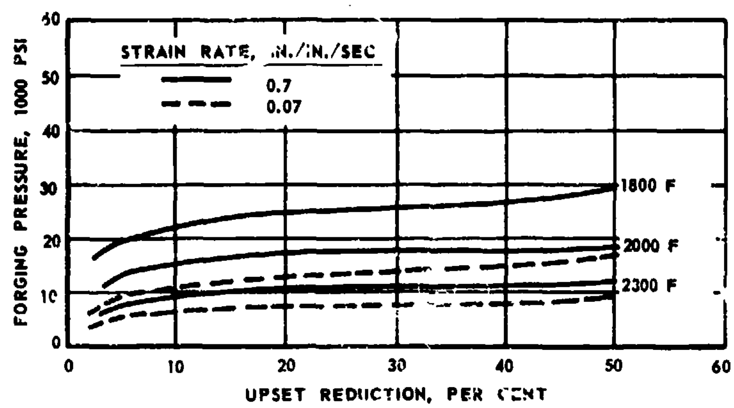


FIGURE 9-10. FORGING PRESSURE FOR AISI 4340 UPSET AT VARIOUS TEMPERATURES AND TWO STRAIN RATES[4]

Similar strain rate effects occur with the low-alloy steels. Figure 9-10 shows the relative forging pressures for AISI 4340 steel at several temperatures over a 10-fold range of strain rates, as determined in upsetting studies on a hydraulic press at Battelle[4]. The effect of strain rate is more evident when this data is compared with upsetting at hammer velocities. Energy measurements for upsetting AISI 4340 steel at 2300 F in a hydraulic press and a Dynapak, tabulated in Table 9-1, show that the specific energy of forging more than doubles over a 100-fold increase in strain rate and triples over a 1000-fold increase in rate.

TABLE 9-1. INFLUENCE OF STRAIN RATE ON ENERGY REQUIRED FOR FORGING AISI 4340 AT THREE TEMPERATURES[4]

Samples 2 in. ϕ x 2 in. long upset in hydraulic press at ram speeds of 0.14 and 1.4 in./sec., and in a Dynapak at nominal ram velocities between 160 and 230 in./sec.

Test Temperature, F	Specific Energy, in-lb/cu in., Required to Accomplish a 50 Per Cent Upset Reduction		
	Nominal Strain Rates, sec^{-1}		
	0.07	0.7	85-100
1800	--	14,300	28,600
2000	7,600	9,700	20,000
2300	4,200	6,200	14,300

Data obtained by Tout and Banning[6] showed similar trends for hot twisting a 0.20 carbon steel. At 2200 F, 20 per cent greater torque was required for twisting bars when the twist rate was increased from 200 to 400 rpm.

INFLUENCE OF FORGING VARIABLES ON MECHANICAL PROPERTIES

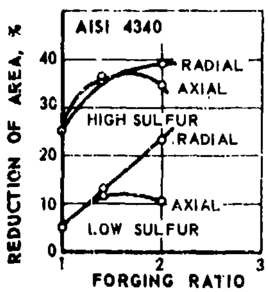
The hardness and strength of steels are controlled by composition and heat treatment, and are not influenced significantly by forging. However, forging usually affects other important mechanical properties, such as ductility, impact strength, and fatigue strength. The generally held reasons for these improvements are:

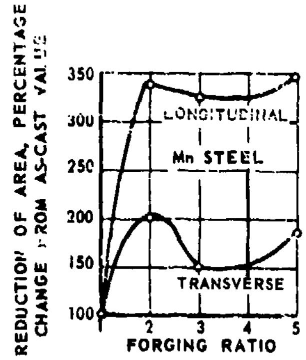
(1) Forging breaks up segregation, heals porosity, and aids homogenization.

(2) Forging produces a fibrous grain structure which enhances mechanical properties parallel to the grain flow.

(3) Forging reduces as-cast grain size.

Figures 9-11 and 9-12 show the typical improvement in ductility and impact strengths of heat-treated steels obtained from increasing reductions. These data show that





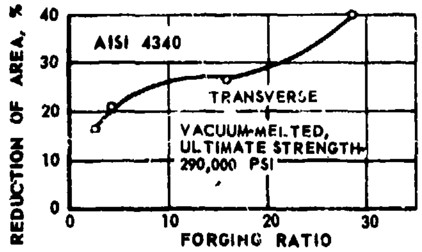


FIGURE 9-11.

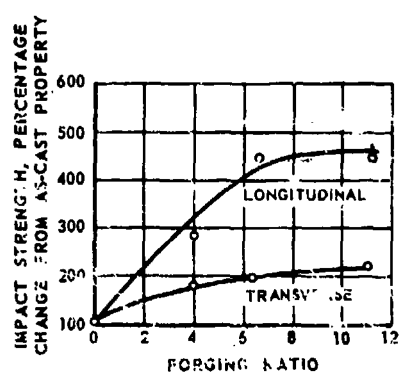
EFFECT OF FORGING RATIO ON REDUCTION OF AREA OF HEAT-TREATED STEELS(7,5,9)

Forging Ratio = $\frac{\text{Final Cross-Sectional Area}}{\text{Original Cross-Sectional Area}}$

FIGURE 9-12.

EFFECT OF HOT-WORKING REDUCTION ON IMPACT STRENGTH OF HEAT-TREATED NICKEL-CHROMIUM STEEL(10)

Forging Ratio = $\frac{\text{Final Cross-Sectional Area}}{\text{Original Cross-Sectional Area}}$



maximum improvement, in each case, occurs in the direction of maximum extension. After a certain amount of hot reduction on these steels, the toughness and ductility properties reach levels beyond which further hot reduction is of little significant value.

Figure 9-13 shows how increasing the forging reduction improves ductility by reducing both porosity and the size of inclusions present in the original as-cast ingot. Porosity is normally eliminated and inclusion size is markedly reduced by reductions of about 50 per cent. Some types of nonmetallic inclusions continue to break up with increasing reduction. Large reductions develop a highly oriented, fibrous structure. As a result, longitudinal ductility continues to improve. Transverse ductility shows little change after porosity is healed and may decrease with very heavy reductions.

It is important to remember that die forgings are made from wrought billets that have received considerable prior reduction and, therefore, porosity is rarely a problem. It is also important to recognize that metal flows in a number of directions during die forging. For example, the metal flow is virtually all in the transverse direction when die forging a rib-and-web shape. Such flow improves transverse ductility at little or no expense of longitudinal ductility. This point is illustrated by the diagram in Figure 9-14. Conceivably, the transverse ductility could equal or surpass longitudinal ductility if forging reductions were great enough and the metal flow was principally in the transverse direction.

Similar effects are observed when upsetting wrought billets. In this case, however, the original longitudinal axis is shortened by upsetting and the lateral displacement of metal is in the radial direction. When upset reductions exceed about 50 per cent, the ductility in the radial direction generally surpasses that in the axial direction as shown in Figure 9-15.

COMMENTARY ON FORGING PRACTICES

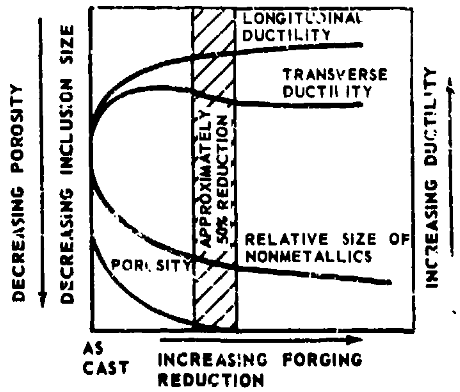
The hundreds of alloy compositions in the carbon and low-alloy steel category exhibit essentially the same forging behavior with the possible exception of the free-machining grades. The most important differences are found in forging-pressure requirements. Several important factors in processing common to the forging of all these steels are discussed below.

Lubrication. A wide variety of lubricants are used for steel forgings. Among the most popular are graphite suspensions, salts, oils, and sawdust. The choice of a lubricant depends on the nature of the die impression and the desired direction of flow. For forging in dies with deep recesses, salts and sawdust are favored because they reduce sticking and prevent excessive lateral flow in the flash region. Oils and graphite suspensions are favored for shallow impressions where considerable lateral flow is required.

Selection of Forging Temperature. The selection of forging temperatures for carbon and low-alloy steels is based on (1) the carbon content, (2) the alloy composition, (3) the temperature range for optimum plasticity, and (4) the amount of reduction. Generally, the maximum temperature allowable by these factors is selected since it insures the lowest possible forging pressures.

FIGURE 9-13.

IMPROVEMENT IN DUCTILITY OF STEELS BY REDUCING BOTH POROSITY AND THE SIZE OF INCLUSIONS DURING FORGING



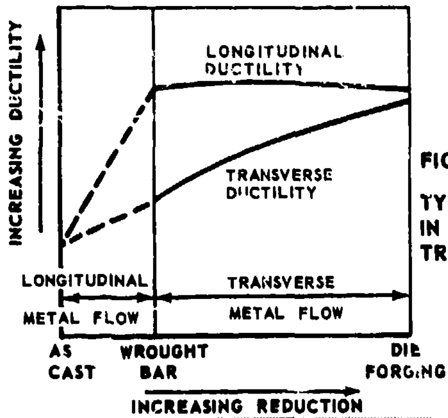
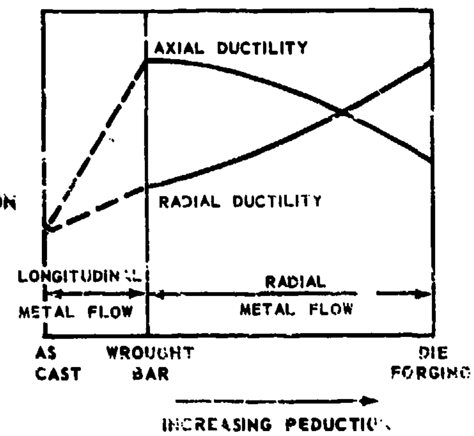


FIGURE 9-14.

TYPICAL INFLUENCE OF LATERAL REDUCTION IN DIE FORGING ON LONGITUDINAL AND TRANSVERSE DUCTILITY OF STEELS

FIGURE 9-15.

TYPICAL INFLUENCE OF UPSET REDUCTION IN DIE FORGING ON AXIAL AND RADIAL DUCTILITY OF STEELS



The upper limiting forging temperatures for steels are influenced most noticeably by carbon content. The maximum forging temperatures for carbon steels with increasing carbon contents are approximately as follows:

AISI Steel Designation	Carbon Content per cent	Maximum Forging Temperature, F
C1010	0.1	2400
C1030	0.3	2350
C1050	0.5	2300
C1080	0.8	2200
C1095	1.0	2150

These forging temperatures are approximately 300 F below the solidus temperature of each composition. Above these temperatures the steels are subject to possible damage by overheating or burning.

Low-alloy steels have solidus temperatures lower than plain carbon steels of comparable carbon contents. Thus, the maximum forging temperatures for the low-alloy steels at each carbon level are generally 50 to 100 lower. In addition, the forging temperatures are usually reduced another 50 to 100 F for parts requiring only small reductions. This provides further insurance that the alloys do not overheat.

Decarburization. Methods for reducing decarburization during forging include the use of controlled-atmosphere furnaces, electroplated coatings, glass coatings, flame-sprayed coatings, hot-dip coatings, and combinations of coatings and atmospheres. The most promising method for preventing decarburization on a common low-alloy steel (AISI 4340) involves the use of an electroplated nickel coating (about 3 mils thick) in combination with a controlled-atmosphere furnace. Experimental studies described by Proffitt[11] indicate that decarburization can be virtually eliminated by using nickel coated steel heated to 2200 F in an atmosphere of nitrogen and atomized pine oil. Another source reported the successful use of hot-dipped aluminum coatings, but this introduces the problem of removing the hard iron-aluminum compound developed on the surface. As yet, no inexpensive method for preventing decarburization is available.

Flakes. Some steels melted in open-hearth and electric furnaces develop internal cracks when cooled from forging. These cracks are called "flakes". It is generally held that flakes are a direct result of excessive occlusion of hydrogen at grain boundaries, lattice imperfections, and other microstructural discontinuities during cooling. Flakes frequently occur in relatively large sections (3 inches or larger) when forgings cool at normal air-cooling rates. However, flaking can be prevented by slowly cooling the forgings in either furnaces or insulating material. For extremely large open-die forgings, it has recently become common practice to reduce hydrogen levels of the starting ingot by vacuum melting, vacuum pouring, or vacuum degassing. All of these techniques reduce or eliminate the problem of flaking and the incidence of delayed brittle failure in service.

Cooling Cracks. Alloys having high hardenability are subject to cooling cracks if cooled too rapidly after forging. This is particularly true of the high-carbon and

3 to 6 per cent chromium grades. Such steels are usually either cooled from forging in insulating material or in furnaces to prevent cracking.

COMMENTARY ON FORGING DESIGN

In Chapter 4 on Forging Design, carbon and low-alloy steels serve as a basis for comparing appropriate designs in steels with those of other metals. Since the alloys in this classification represent hundreds of alloy combinations in an even greater variety of forgeable shapes and sizes, it is impossible to formulate hard and fast rules about forging design.

However, there are practical limits to forging size, shape, and design in each of the commonly used types of forging equipment. Table 9-2 gives practical limits on these design factors for each of four types of forging equipment. In general, faster equipment is capable of producing thinner sections. The minimum thickness and tolerances increase with increasing forging size.

TABLE 9-2. SOME PRACTICAL SIZE AND SHAPE LIMITATIONS IMPOSED ON CARBON AND LOW-ALLOY STEEL FORGINGS BY EQUIPMENT SIZE AND FORGING RATE[4]

	Hydraulic Presses		Mechanical Presses	Drop and Counterblow Hammers	High-Energy-Rate Machines
	Conventional	Heavy[a]			
		Equipment Characteristics			
Maximum Equipment Size	20,000 tons	50,000 tons	8,000 tons	150,000 lb	225,000 ft-lb
Forging Action	Single stroke	Single stroke	Single stroke	Repetitive strokes	Single stroke
Range of Forging Rates, inches	Up to 2	Up to 1/2	Up to 10	150-250	200-2000
		Forging Size and Shape Limitations			
Weight Limitation	About 2,000 lb	--	Up to about 50 lb	--	About 15 lb
Maximum Dimension, inches	About 10 ft	About 20 ft	About 2 ft	About 15 ft	10 inches
Practical Minimum Thickness, inches	3/8	1/2	1/8	3/16	3/32
Practical Minimum Thickness Tolerance, inch, For Typical Disk Forgings					
5-in. diameter	±0.030	±0.030	±0.020	±0.0[illegible]0	±0.010
10-in. diameter	±0.050	±0.050	±0.030	±0.045	--
20-in. diameter	±0.090	±0.090	--	±0.070	--

(a) Heavy presses are the 3[illegible] 000-ton and 50,000-ton presses built under the Air Force Heavy Press Program.

REFERENCES

(1) Ihrig, H. K., "The Effect of Various Elements on the Hot Workability of Steels", Trans. AIME, 167, 749-777 (1946).

(2) Clark, C. L., and Russ, J. J., "A Laboratory Evaluation of the Hot Working Characteristic of Metals", Trans. AIME, 167, 736-748 (1945).

(3) Anderson, C. T., Kimball, R. W., and Cattoir, F. R., "Effect of Various Elements on the Hot Working Characteristics and Physical Properties of Fe-C Alloys", J. Metals, 5 (4), 525-529 (April, 1953).

(4) Henning, H. J., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", Technical Documentary Report No. ML-TDR-64-95 on Contract AF AF 33(600)-42963, Battelle Memorial Institute (March, 1964).

(5) Alder, J. F., and Phillips, V. A., "The Effect of Strain Rate and Temperature on the Resistance of Aluminum, Copper, and Steel to Compression", J. Inst. of Metals, 83, 80-86 (1954-1955).

(6) Tout, C. S., Jr., and Banning, L. H., "A Rapid Twist Test for Determining Hot Forming Temperatures of Steels", Report of Investigation 5928, U. S. Dept. of Interior, Bureau of Mines (1962).

(7) Boulger, F. W., et al., "A Study on Possible Methods for Improving Forging and Extruding Process for Ferrous and Nonferrous Materials", Final Engineering Report on Contract No. AF 33(600)-26272, Battelle Memorial Institute (June 30, 1957).

(8) Maurer, E., and Korshan, H., "Contribution to the Knowledge of the Mechanical Properties of Very Large Forgings", Stahl u. Eisen, 53, 209 (March, 1953).

(9) Sprague, L. E., "The Effects of Vacuum Melting on the Fabrication and Mechanical Properties of Forgings", Steel Improvement and Forge Company, Cleveland, Ohio (June, 1960).

(10) Voss, H., "Relations Between Primary Structure, Reduction in Forging, and Mechanical Properties of Two Structural Steels", Arch. Eisenhüttenw., 7, 403-406 (1933-1934).

(11) Proffitt, C. E., "Close Tolerance Steel Forging Development", Boeing Airplane Company, Technical Engineering Reports on Air Force Contract AF 33(600)-36659.

CHAPTER 10

MARTENSITIC STAINLESS STEELS

Crystalline Structure:	Austenite – face-centered cubic Ferrite – body-centered cubic Martensite – tetragonal
Phase Changes:	Austenite transforms to ferrite plus carbide over a range of temperatures below 1500 F Austenite transforms to martensite in the range of about 300-500 F on rapid cooling Austenite transforms to delta ferrite on heating low-carbon alloys above 2000 F
Number of Phases:	One phase (austenite) at forging temperature
Liquidus Temperature Range:	2650 to 2750 F
Solidus Temperature Range:	2500 to 2700 F
Reactions When Heated In Air:	Oxidation and possible decarburization depending on temperature and carbon content.

ALLOYS AND FORMS AVAILABLE

Nominal compositions for several martensitic stainless steels are given in Table 10-1. The sizes of ingots, billets, and bars available for forging represent a range essentially as broad as that for carbon and low-alloy steels. Conventional procedures are used for melting, cogging, and rolling.

TABLE 10-1. NOMINAL COMPOSITIONS OF SEVERAL MARTENSITIC STAINLESS STEELS

		Nominal Chemical Composition, per cent						
Alloy	Specifications	C	Mn	Si	Cr	Ni	Mo	Others
410	AMS 5613	0.10	0.7	0.3	12.5	--	--	--
414	AMS 5615	0.12	0.7	0.3	12.5	1.7	--	--
Greek Ascoloy	AMS 5616	0.18	0.3	0.2	13.0	2.0	[illegible]	W 3.0
416	--	0.10	0.9	0.5	13.0	--	--	S 0.15 min
420	--	0.20	0.7	0.5	13.0	--	--	--
440A		0.65	0.7	0.5	17.0	--	0.5	--
440B		0.85						
440C		1.10						
Lapelloy C		0.20	0.9	0.3	11.5	0.3	2.7	N 0.05 Cu 2.0

FORGING BEHAVIOR

The high-chromium, martensitic steels have forging characteristics similar to low-alloy steels. Because of the higher chromium contents, however, forging load requirements are about 30 to 50 per cent higher. At temperatures above 2000 F, certain of the alloys transform partly to delta ferrite which reduces forgeability. For this reason, maximum forging temperatures are generally 100 F to 300 F lower than for low-alloy steels.

The austenite-to-ferrite transformation has an important influence on forgeability and forging temperature. For this reason, Bloom, et al[1] studied the influences of temperature on twist ductility for stainless grades having chromium contents between 12 and 26 per cent. They concluded that the delta-ferrite formation temperature decreases with increasing chromium contents and that small amounts of delta ferrite reduce forgeability significantly. The studies also showed that as the amount of delta ferrite increases above about 15 per cent, forgeability improves gradually until the structure becomes entirely ferritic where twist ductility increases sharply. The temperature range producing a totally ferritic structure in the martensitic stainless steels, however, is too close to the melting range to serve as a practical forging temperature.

Figure 10-1 shows that the twist ductility of AISI 410 stainless steel at 2100 F is fairly high but decreases rapidly at 2200 F to a level comparable to the high-sulfur 416 stainless grade.

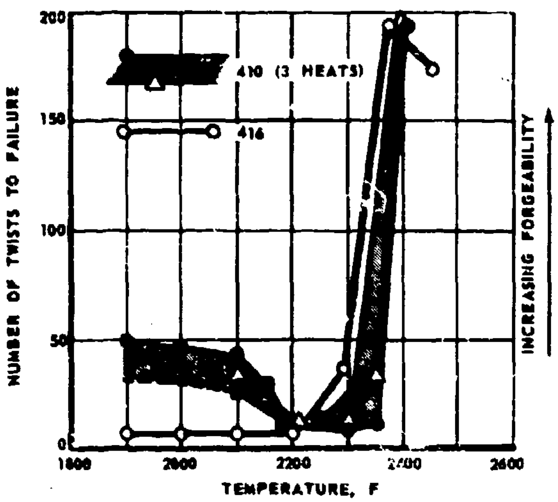


FIGURE 10-1.

INFLUENCE OF TEMPERATURE ON TWIST DUCTILITY AND FORGEABILITY OF 410 AND 416 STAINLESS STEELS

After Bloom, et al.

Figure 10-2 gives twist data for two heats of AISI 420 stainless showing typical heat-to-heat variation. Much less effect of temperature on twist ductility is evident for this stainless steel than for the 410 grade steel as was shown in Figure 10-1.

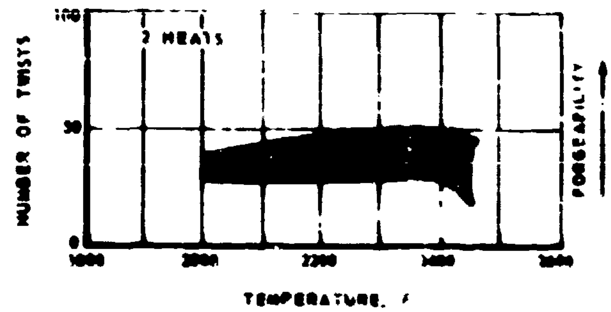


FIGURE 10-2. INFLUENCE OF TEMPERATURE ON TWIST DUCTILITY AND FORGEABILITY OF 420 STAINLESS STEEL

After [illegible][2]

Figure 10-3 compares twist data for several martensitic stainless steels with that for AISI 4340 steel. The twist ductility values for the AISI 410 and 420 grades drop sharply above 2200 F and 2400 F, respectively. The 440A and 440C grades do not exhibit such drastic changes with temperature, but these grades also do not exhibit high levels of ductility. Relative torque values at temperatures corresponding to maximum twist ductility are compared with AISI 4340 below:

Alloy	Recommended Forging Temperature, F	Torque Relative to AISI 4340
AISI 4340	2300	1.0
AISI 410	2150	1.5
AISI 420	2200	1.4
AISI 440A	2100	2.2
AISI 440C	2050	2.0

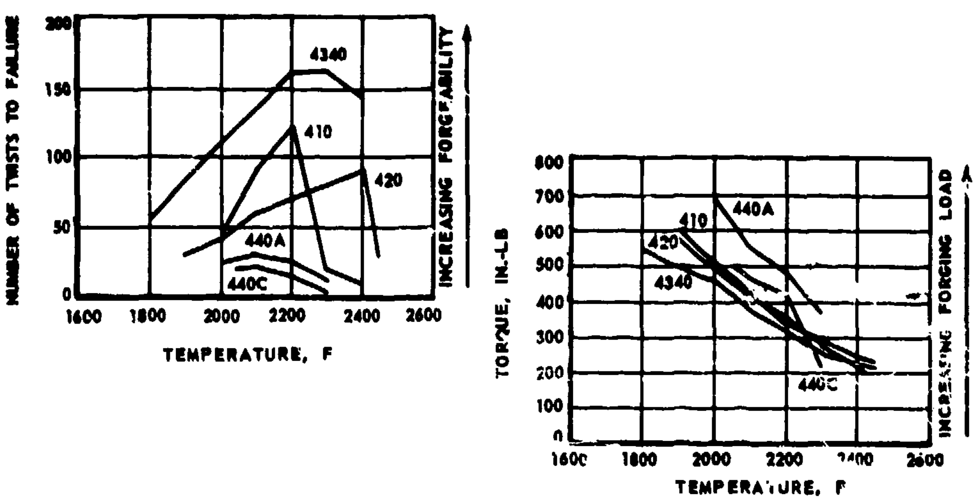


FIGURE 10-3. FORGING CHARACTERISTICS OF SEVERAL MARTENSITIC STAINLESS STEELS COMPARED TO 4340 STEEL, AS MEASURED IN HOT-TWIST TESTS

After Clark and Russ[3]

Such comparisons with the more familiar low-alloy steels are useful for estimating the forces required to forge these alloys. These forging temperatures agree well with

those used in forging practice. The data indicate that the greatest forging-load requirements would be expected for the AISI 440 alloys.

COMMENTARY ON FORGING PRACTICE AND DESIGN

While there are comparatively few quantitative data on the forging characteristics of the martensitic stainless grades, the important factors are well understood. The most serious problem in the forging of martensitic stainless steels is the presence of delta ferrite. This phase significantly reduces the transverse ductility of these steels. This is particularly true of the ductility in the short transverse direction. In some heavy forgings, the ductility of martensitic stainless steels has been of the order of 1-2 per cent in this direction, while in the longitudinal direction the ductility may be as high as 15 per cent.

Selection of Forging Temperatures. In general, maximum forging temperatures are determined by the first appearance of the delta-ferrite phase on heating. In practice, these temperatures are readily determined by conducting upset forging trials at temperatures between 2000 F and 2300 F. The first appearance of the phase on heating correlates with the onset of edge cracking.

A chart observed at one company showed that variations in the forgeability of 410 stainless steel could be predicted, to some extent, from composition. Although specific data were not made available, the chart showed that maximum forging temperatures decrease with increasing chromium and with increasing carbon contents. A schematic representation of this type of chart is given in Figure 10-4. Accurate charts of this kind prove useful for predicting forgeability of the various martensitic stainless steels from chemical composition.

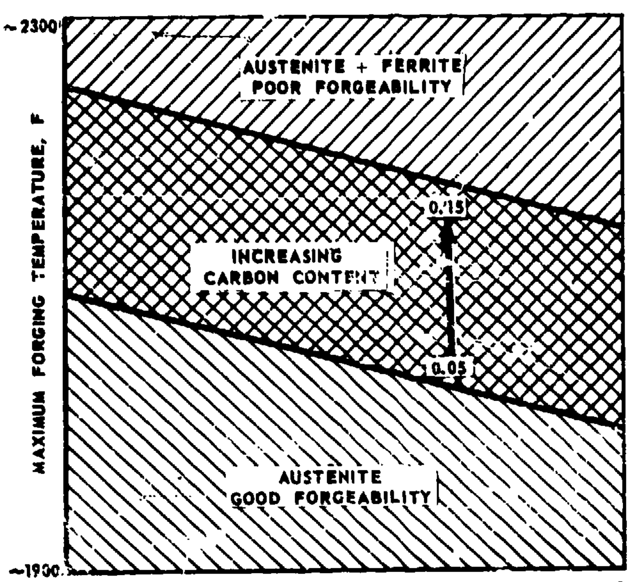


FIGURE 10-4. TYPICAL INFLUENCE OF CHROMIUM AND CARBON VARIATIONS ON THE MAXIMUM FORGING TEMPERATURE FOR MARTENSITIC STAINLESS STEELS

Example based on 410 stainless.

It is important to recognize that the carbon content near the billet surface may be lowered by decarburization. This causes delta ferrite to form at unexpectedly low temperatures and leads to poor forgeability. To minimize decarburization, billets should be heated for the minimum possible times before forging.

Forging Cycle. Forging cycles for the martensitic stainless steels are essentially the same as those used for carbon and low-alloy steels. Because greater pressures are required, larger forging equipment is necessary for otherwise similar parts. When forging with large reductions, workpiece temperatures may increase enough to exceed the delta-ferrite transformation temperature and cause cracking. This can be avoided either by lowering the forging temperature or by forging slowly.

Based on comments from representatives of the forging industry and also on hot-twist test results, the following data are given for comparing both forging loads and forging temperatures for various stainless grades (the relative forging loads are based on a rating of 1.0 for AISI 4340 steel at 2300 F):

Alloy	Forging Temperature, F	Relative Forging Load
AISI 4340	2300	1.0
AISI 410	2150	1.2-1.5
AISI 414	2150	1.3-1.5
AISI 416	2150	1.3-1.5
Greek Ascoloy	2200	1.5-1.7
AISI 420	2200	1.1-1.2
AISI 440A	2100	2.0
AISI 440B	2100	2.0
AISI 440C	2050	2.0
Lapelloy C	2250	2.0

Lubrication. Lubrication techniques are essentially the same as those for carbon and low-alloy steels.

Cooling From Forging. Because the martensitic stainless steels are characterized by high hardenability, they are subject to thermal cracking when cooled from forging temperatures. Normal practice consists of placing hot forgings in insulating materials to provide for slow cooling. For parts that have either heavy sections or large variations in section, it is often desirable to charge the forged parts directly into an annealing furnace.

In particular, the higher-carbon grades, 440A, 440B, and 440C, and the modified 420 types such as Greek Ascoloy and Lapelloy C must be slow cooled carefully after forging. These steels often require furnace-controlled interrupted cooling cycles to insure against cracks. Suitable cycles consist of air cooling the forgings to temperatures where the martensite transformation is partially complete (between 300 and 500 F), then reheating in a furnace at a temperature of about 1200 F before finally cooling to room temperature. This procedure also prevents the formation of excessive grain-boundary carbides that sometimes develop during continuous slow cooling.

Heat Treatment. Martensitic stainless steel forgings may be supplied either for high-strength, wear-resistant applications, or for long-time service at elevated temperatures up to 1100 F. For high-strength applications, full heat treatment is required. This entails austenitizing at 1750 to 1850 F depending upon the type of steel, cooling either in air or oil to room temperature, and subsequently tempering at 600 to 800 F.

Tempering in the range from 650 to 1000 F should be avoided because secondary hardening occurs at such temperatures. Secondary hardening leads to reduced ductility and poor impact resistance.

For high-temperature applications, air cooling from the austenitizing temperature followed by tempering at 1050 to 1250 F is common practice. Often double tempering is employed to reduce the likelihood of subsequent transformation of retained austenite.

Design. Design principles applying to carbon and low-alloy steels are applicable to these stainless grades. Broader dimensional tolerances are ordinarily required, however, owing to the greater forging-pressure requirements.

REFERENCES

(1) Bloom, F. K., Clarke, W. C., Jr., and Jennings, P. A., "Relation of Structure of Stainless Steel to Hot Ductility", Metal Progress, 250-256 (February, 1951).

(2) Ihrig, H. K., "The Effect of Various Elements on the Hot Workability of Steels", Trans. AIME, 167, 749-777 (1946).

(3) Clark, C. L., and Russ, J. J., "A Laboratory Evaluation of the Hot Working Characteristics of Metals", Trans. AIME, 167, 736-748 (1946).

CHAPTER 11

AUSTENITIC STAINLESS STEELS

Crystalline Structure:	Austenite – Face-centered cubic
Phase Changes:	None on cooling, but steels must be cooled rapidly from 1500 F to 900 F to avoid intergranular precipitation of chromium carbide. Austenite transforms to delta ferrite (b.c.c.) on heating lower carbon alloys above 2200 F
Number of Phases:	One phase (austenite) at forging temperatures
Liquidus Temperature Range:	2500 to 2600 F
Solidus Temperature Range:	2450 to 2500 F
Reactions When Heated In Air:	Oxidizes slowly

ALLOYS AND FORMS AVAILABLE

Nominal compositions and forgeability data for several austenitic stainless steels are given in Table 11-1. Ingots and billets are available in virtually any of the size

TABLE 11-1. NOMINAL COMPOSITIONS AND FORGING CHARACTERISTICS OF SEVERAL AUSTENITIC STAINLESS STEELS

	AISI Type					
	303	304	310	316	321	347
Nominal Composition, %	0.1C 18.0Cr 9.0Ni 0.15S	0.05C 19.0Cr 9.0Ni	0.15C 25.0Cr 20.0Ni	0.05C 17.0Cr 12.0Ni 2.5Mo	0.05C 18.0Cr 10.0Ni (5xC)Ti	0.05C 18.0Cr 11.0Ni (10xC)Cb+Ta
Solidus Temp, F	2400	2550	2500	2500	2500	2500
Austenite to Delta Ferrite, F	--	--	2400	2400	2350	--
Recrystallization Temp, F	Between 1550 F and 1700 F depending on prior cold-working conditions					
Forging Temp, F						
Range	1700-2300	1600-2400	1700-2400	1700-2400	1[illegible]-2350	1600-2350
Recommended	2150	2250	2250	2250	2200	2200
Relative Forgeability	Poor	Good	Good	Good	Fair	Fair to good

ranges that are available in carbon and low-alloy steel. The alloys are normally melted by electric-furnace techniques. They are either forged or hot rolled to billet and supplied in the annealed condition for forging.

FORGING BEHAVIOR

By comparison with carbon and low-alloy steels, the austenitic stainless steels require greater forging pressures. They are generally more difficult to forge, they exhibit work-hardening behavior at higher temperatures, and their forging-temperature ranges are narrower. For example, AISI 303 stainless steel contains sulfur for machinability purposes and therefore will contain sulfide inclusions. The presence of these inclusions greatly impairs the hot ductility and forging characteristics. Grades AISI 321 and 347, which contain titanium and columbium, respectively, are prone to contain carbonitride stringers which also can lead to forging difficulties, although these two steels are more readily forgeable than AISI 303.

Hot-twist test data are useful for comparing the forgeabilities of these stainless grades. Figure 11-1 compares the twist ductilities and torque requirements for two stainless grades with those for AISI 1040 and AISI 4340. These data(1) show that both stainless grades (AISI 303 and AISI 304) have about equal torque values that are 25-50 per cent higher than the carbon and low-alloy steels. However, the twist ductility for the AISI 303 grade is much lower than that for the AISI 304 grade primarily because of the high sulfur content.

Figure 11-2 gives twist and torque curves for several other austenitic stainless steels.(1) Again these curves show that, although the twist ductilities differ widely, the torque values are quite similar for all of the alloys tested.

Correlations of hot-twist data on these steels with forging practice are quite good. In practice, the forgeabilities of Types 321 and 347 are quite variable because they frequently contain carbides that promote rupturing.

Ihrig(2) also studied the hot-twist behavior of austenitic stainless steels and observed forgeabilities in general agreement with Clark and Russ.(1) Ihrig compared the twist data with those for actual tube-piercing operations and found that heats exhibiting the lowest twist ductility (number of twists to failure) showed the highest incidence of ruptures during piercing. Distinct correlations were found with different heats of the 304 and 321 grades. Heats of Type 304 exhibiting 95, 51, and 35 twists to failure showed excellent-to-poor performance in that order. Similarly, heats of Type 321 exhibiting 86, 60, and 42 twists to failure showed progressively poorer piercing behavior.

Data useful for indicating relative forgeability also have been determined by Nippes, et al.(3) Small specimens (0.25" dia. x 4") of several alloys were tested in tension at elevated temperatures and at strain rates estimated to be about 1.0 in./in./sec. Figure 11-3 summarizes the influence of temperature on the hot ductility (reduction of area) for five steels. Since the specimens were heated rapidly (in less than 5 seconds), the values are not very representative of actual forging conditions. However, the curves show sharp drops in ductility at critical temperatures for each alloy.

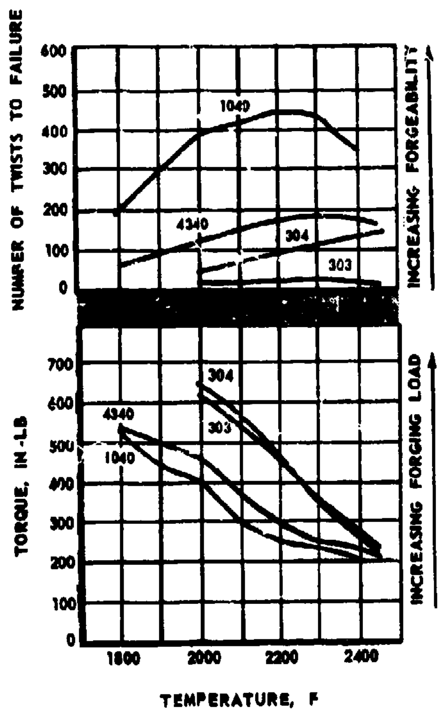


FIGURE 11-1.

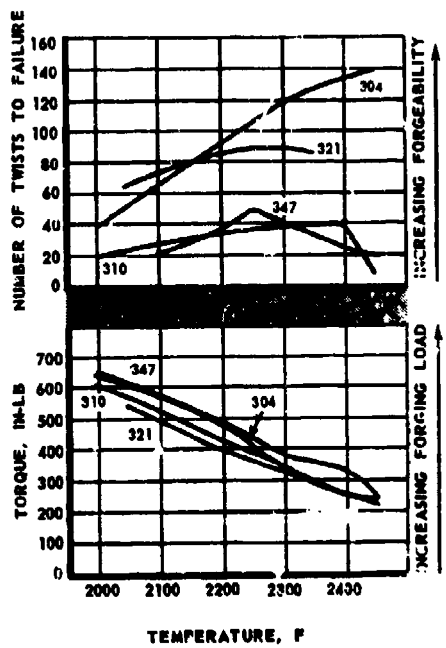
COMPARISON OF FORGING CHARACTERISTICS OF TYPICAL AUSTENITIC STAINLESS STEELS WITH CARBON AND LOW-ALLOY STEELS, BASED ON HOT-TWIST DATA

After Clark and Russ(1)

FIGURE 11-2.

INFLUENCE OF TEMPERATURE ON FORGING CHARACTERISTICS OF SEVERAL AUSTENITIC STAINLESS STEELS, BASED ON HOT-TWIST TESTS

After Clark and Russ(1)



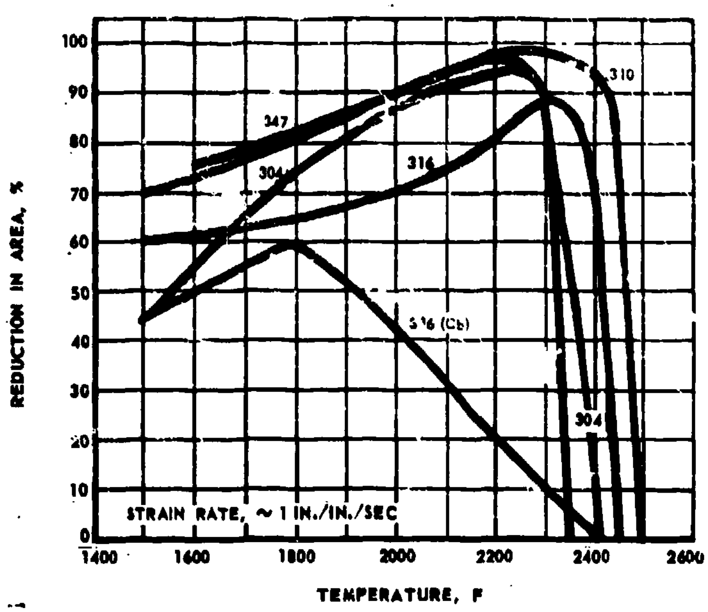


FIGURE 11-3. HOT DUCTILITY OF SEVERAL AUSTENITIC STAINLESS STEELS DETERMINED BY RAPID TENSILE TESTS

After Nippes, et al.[3]

Upset forging studies conducted at Battelle[4] demonstrate that austenitic stainless steels can require increasing forging pressures with increasing reductions even at temperatures thought to be true hot-working temperatures. Figure 11-4 shows how forging pressures of Type 304 stainless steel are influenced by increasing reductions at each of four temperatures. Figure 11-5 compares the variation of forging pressure with temperature and forging reduction between Type 304 and AISI 1020 carbon steel. At 1800 F, 1020 carbon steel requires pressures from 21,000 to 25,000 psi over a range of upset reductions from 5 to 50 per cent. The forging pressure for AISI 304 stainless steel, however, increases from 33,000 to 57,000 psi under the same conditions.

Bloom, et al.,[5] studied the influence of structural changes on the hot-twist ductility of austenitic stainless steels. Tests were conducted on austenitic steels with similar chromium and nickel contents to show how small variations in composition influence the critical temperature at which the austenite phase transforms partly to the delta ferrite phase. Figure 11-6 shows the amount of delta ferrite present and the twist ductility at temperatures in the range 2000-2500 F for four steels. This series of graphs also shows how titanium and molybdenum promote the formation of delta ferrite in austenitic stainless steels. For each steel, the first appearance of the delta ferrite phase coincided with a drop in twist ductility.

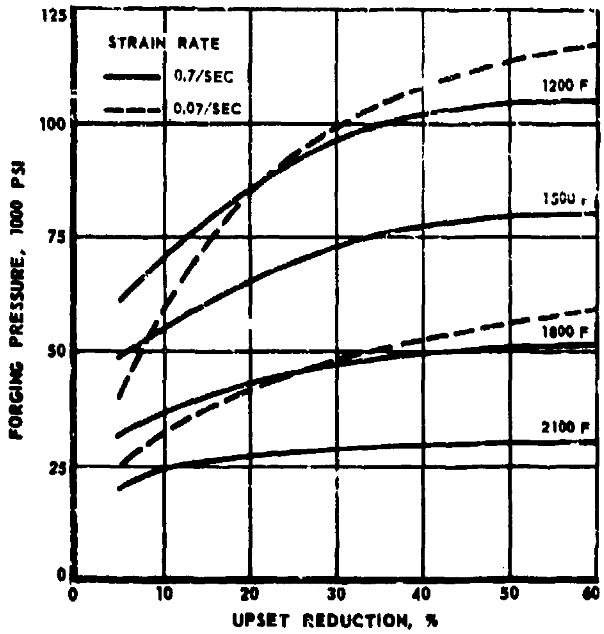


FIGURE 11-4.

FORGING-PRESSURE CURVES FOR TYPE 304 STAINLESS STEEL UPSET AT VARIOUS TEMPERATURES AND STRAIN RATES

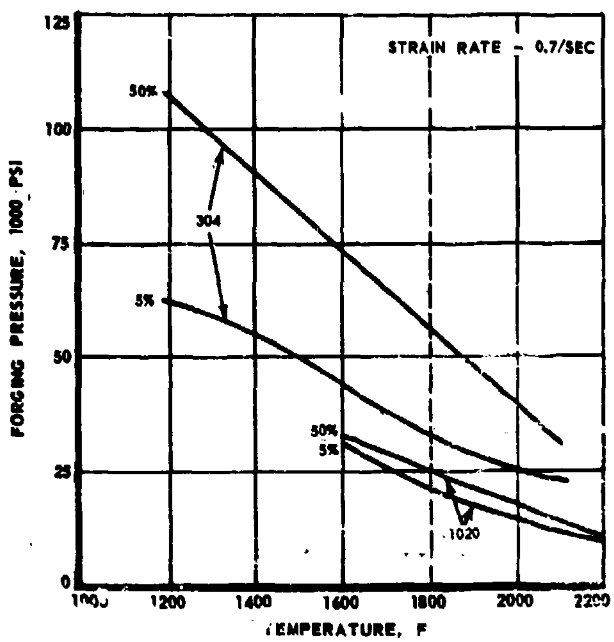


FIGURE 11-5.

COMPARISON OF FORGING-PRESSURE REQUIREMENTS FOR TYPE 304 STAINLESS AND 1020 CARBON STEELS AT SEVERAL TEMPERATURES

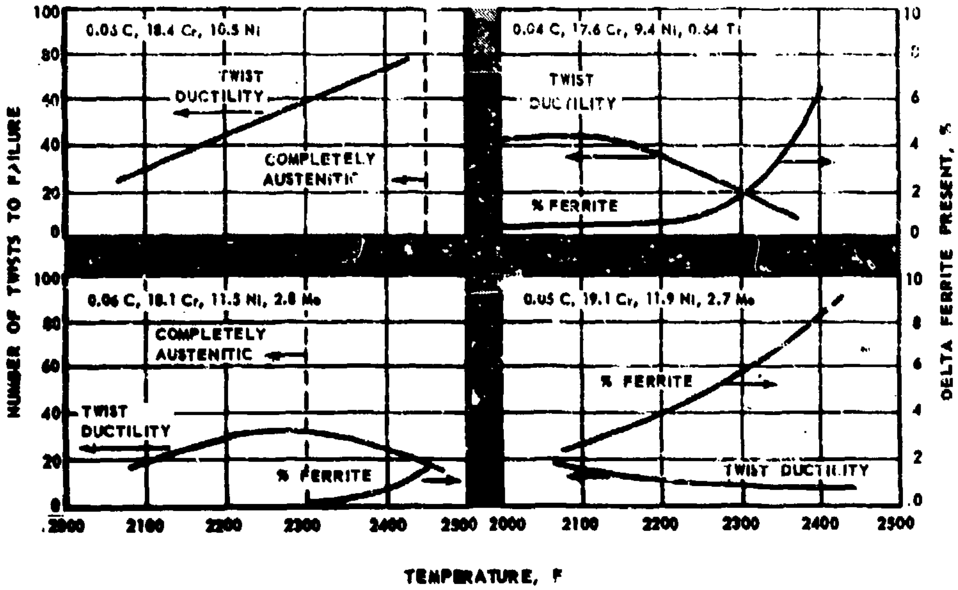


FIGURE 11-6.

MICROSTRUCTURE AND TWIST DUCTILITY OF SEVERAL AUSTENITIC STAINLESS STEELS WITH SIMILAR CHROMIUM AND NICKEL CONTENTS AT TEMPERATURES FROM 2000 F TO 2500 F

COMMENTARY ON FORGING PRACTICES

Austenitic stainless steels possess higher hot strengths than carbon, low-alloy, or even martensitic stainless steels and, in fact, exhibit work hardening at relatively high temperatures. Maximum forging temperatures up to about 2300 F are possible for the 18-8 types, but this limit is considerably lower for those grades that tend to form delta ferrite which, in turn, seriously impairs forgeability. Minimum forging temperatures must generally be above about 1700 F to avoid cracking and hot tearing. It is important, therefore, to select the forging temperatures carefully and exercise equal care during forging with regard to the intensity of hammer blows, the amount of reduction between reheatings, the final reduction, etc.

Selection of Forging Temperatures. Maximum and minimum forging temperatures for austenitic stainless steels, from the standpoint of forgeability, can be readily determined from hot-twist data. However, the selection of optimum temperatures must consider the final properties of the forging. Parts receiving small reductions are subject to excessive grain growth if forged at the maximum temperatures. It is common practice, therefore, to adjust forging temperatures downward for forging

operations requiring small amounts of reduction. For example, the typical forging temperatures for AISI 304 stainless steel for various operations would be as follows:

	Forging Temperature, F
Severe reductions (ingot breakdown, roll forging, drawing, blocking, back extrusion, etc.)	2300
Moderate reductions (finish forging, upsetting, etc.)	2200
Slight reductions (coining, restriking, end upsetting, etc.)	2050

Similar adjustments are made for the other austenitic grades. The temperatures used for slight reductions rarely exceed 2100 F regardless of the alloy being forged.

In addition, it is essential to select forging temperatures which will not produce delta ferrite. Delta ferrite, in addition to giving rise to edge-cracking in forging, will reduce the transverse ductility (particularly in the thickness direction) of finished forgings. Furthermore, many austenitic steels are marketed on the basis of being completely austenitic.

Lubrication. Since austenitic stainless steels do not form much scale during heating for forging, they have a tendency to seize in dies during forging. For this reason, the choice of lubricants is more critical than for carbon, low-alloy, and martensitic stainless steels. Most common die-forging lubricants are oil-base suspensions of graphite, but molten glasses are often used for back extrusion

One technique for reducing seizing consists of tempering the forging dies after the final machining step. The oxide layer formed on the dies serves as a parting agent and provides a bond for the graphite lubricants. Carburization from graphitic lubricants sometimes presents a problem. This problem can be avoided by thoroughly cleaning the forgings before reheating.

Furnace Atmospheres. The higher-nickel austenitic stainless steels are susceptible to attack by sulfur. If sulfur-rich fuels must be used, it is important to maintain a slightly oxidizing atmosphere. In addition, the low-carbon austenitic types (AISI 304, 304L, 316, 316L, etc.) will carburize readily. By maintaining oxidizing atmospheres, carburization is not likely to occur.

An air atmosphere is recommended for all heating of austenitic stainless steels. Scale will form but this can readily be removed by acid pickling in a hot nitric-hydrofluoric acid solution. Scale formed under reducing furnace conditions is extremely difficult to remove by pickling and necessitates a twofold pickling process in which the scale is first oxidized with sulfuric acid and then removed with the normal nitric-hydrofluoric acid pickling solution.

<u>Forging Equipment Requirements</u>. Neither forging pressure nor the forgeability of austenitic stainless steels are influenced significantly by deformation rates. Hammers and other fast-acting equipment are preferred, but only because they minimize the problems associated with heat transfer. Two to three times as much energy is required for forging the 300-series stainless steels than for forging carbon and low-alloy steels. In hammer forging, this means either larger hammer capacity or more hammer blows per forging. In most cases, forging companies will compromise with about 50 per cent greater hammer capacity and 50-100 per cent more hammer blows. Press forgers figure the maximum forging plan areas at about half that for the carbon and low-alloy steels for any given press size. For the hot-cold-working alloys, equipment and requirements vary depending on the strength desired.

<u>Heat Treatment</u>. The austenitic stainless steels are noted for their good strength at elevated temperatures and for their good resistance to corrosion. Because a large proportion of these steels are employed in applications where good corrosion resistance is required, it is mandatory that these materials be supplied in a condition best suited for resistance to corrosive attack.

When slowly cooled from the forging temperature, those steels which are not stabilized (do not contain columbium or titanium) will contain intergranular chromium carbides and will be subject to intergranular corrosion in certain media. To prevent this, the forged parts must be annealed at about 1950 F and rapidly cooled through the temperature range from 1500 to 900 F. This will retain the undesirable chromium carbides in solid solution and reduce subsequent intergranular attack.

The stabilized grades (Types 321 and 347) also must be annealed after forging to insure that all of the carbon is tied up as titanium or columbium carbides. This is accomplished by heating these stabilized grades at 1600-1650 F, a temperature at which the chromium carbides are dissolved but the titanium or columbium carbides are precipitated. After such annealing of Types 321 and 347, slow cooling to room temperature will not impair the corrosion resistance of the steels.

REFERENCES

(1) Clark, C. L., and Russ, J. J., "A Laboratory Evaluation of the Hot Working Characteristics of Metals", Trans. AIME, <u>167</u>, 736-748 (1946).

(2) Ihrig, H. K., "A Quantitative Hot Workability Test for Metals", <u>Iron Age</u>, 85 (April, 1944).

(3) Nippes, E. F., Swage, W. F., Grotke, G., "Further Studies on the Hot Ductility of High Temperature Alloys", Report No. 33 Bulletin Series. Welding Research Council of the Engineering Foundation, New York, N. Y. (February, 1957).

(4) Henning, H. J., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", Technical Documentary Report No. ML-TDR-64-95, Contract AF 33(600)-42963, Battelle Memorial Institute (March, 1964).

(5) Bloom, F. K., Clarke, W. C., Jr., and Jennings, P. A., "Relation of Structure of Stainless Steel to Hot Ductility", <u>Metal Progress</u>, 250 (February, 1951).

CHAPTER 12

PRECIPITATION-HARDENABLE STAINLESS STEELS

Crystalline Structure:	Austenite – Face-centered cubic Ferrite – Body-centered cubic
Phase Changes:	Austenite stable to temperatures as low as -100 F; under certain conditions, austenite will transform to martensite on cooling. On heating above 2000 F austenite transforms partly to delta ferrite.
Number of Phases:	Normally one (austenite) at forging temperature. Certain alloys contain delta ferrite phase (up to 20 per cent) over entire forging-temperature range
Solidus Temperature Range:	2450 to 2650 F
Reactions When Heated in Air:	Oxidation

ALLOYS AND FORMS AVAILABLE

Precipitation-hardenable stainless steels have been developed by making certain alloy additions to martensitic, semiaustenitic, and austenitic alloys.[1-4] Thus, the martensitic precipitation-hardenable stainless steels are similar to, but distinct from, the martensitic stainless steels. The austenitic precipitation-hardenable stainless steels are likewise different from the austenitic stainless steels. The austenitic precipitation-hardenable stainless steels are known for their relatively high strength in the range 900 to 1300 F, and are often referred to as iron-base superalloys. These steels are described under that category in Chapter 15.

Nominal compositions of several semiaustenitic and martensitic precipitation-hardenable stainless steels are given in Table 12-1. Figure 12-1 shows a schematic constitution diagram for stainless steels. The two vertical lines superimposed on the curve represent ranges of composition characteristic of these steels. The location of this band shifts with the composition, as indicated by the arrows on the graph.

All of the semiaustenitic stainless steels are either precipitation hardenable or classed as such, and they combine some of the features of both the martensitic and the austenitic stainless steels. As annealed, these steels are austenitic, soft, and readily cold formed. As fully hardened, they are martensitic and attain strength properties somewhat higher than the 12-14 per cent chromium martensitic stainless steels (e.g., Type 410).

TABLE 12-1. NOMINAL COMPOSITIONS OF SEVERAL PRECIPITATION-HARDENABLE STAINLESS STEELS

Alloy	Nominal Composition, per cent										
	C	Mn	Si	Cr	Ni	Mo	Ti	Al	V	Cu	Others	Fe
					Semiaustenitic							
AM-350	0.10	0.75	0.35	16.0	4.25	2.75	--	--	--	--	0.10 N	Bal
AM-355	0.13	0.85	0.35	15.5	4.25	2.75	--	--	--	--	0.12 N	Bal
17-7PH	0.07	0.70	0.40	17.0	7.0	-	--	1.15	--	--	-	Bal
PH 15-7 Mo	0.07	0.70	0.40	15.0	7.0	2.25	--	1.15	--	--	--	Bal
					Martensitic							
17-4PH	0.04	[illegible]	0.50	16.5	4.25	0.50	--	--	--	3.6	0.25 Cb	Bal
Stainless W	0.07	0.60	0.50	17.0	7.0	--	0.6	0.2	--	--	0.1 N	Bal

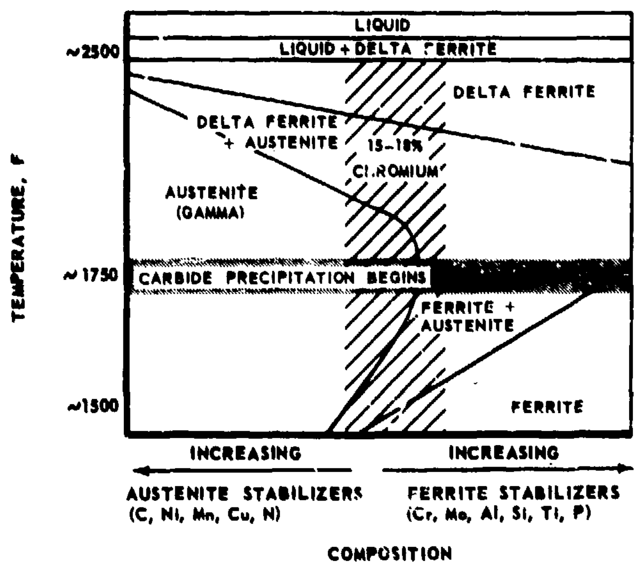


FIGURE 12-1. SCHEMATIC CONSTITUTION DIAGRAM FOR STAINLESS STEELS SHOWING TYPICAL RELATIONSHIP BETWEEN COMPOSITION AND STRUCTURE

Careful control of composition and heat-treating procedures is required to obtain this combination of characteristics. The balance between the austenite stabilizers (nickel, carbon, nitrogen, and manganese) and the ferrite stabilizers (chromium, silicon, molybdenum, and aluminum) is adjusted so that after normal annealing at 1900 to 1950 F, the steel is predominantly austenitic at room temperature. By reannealing at lower temperatures (1400 to 1750 F), chromium carbides are precipitated rendering the austenitic matrix less stable and raising the M_s temperature (the temperature of the start of martensite formation). After treating at 1400 F and cooling to room temperature, the structure is completely martensitic. After annealing at 1750 F, the M_s temperature is raised to about 0 F. Refrigeration at -100 F will then produce a martensitic structure in such materials. When the martensitic structure is obtained, the steel is in a condition whereby precipitation hardening will occur in the temperature range from 850 to 1050 F. This will further increase the strength of the steel. In the 17-7PH and PH 15-7 Mo steels, the precipitating phase is aluminum rich, while in the AM 350 and AM 355 alloys it is believed to be a nitride.

The martensitic precipitation-hardening stainless steels transform to a relatively "soft" martensite (R_c30) at temperatures in the range of 200 to 300 F during cooling from forging temperatures. Additional strengthening (R_c43) is derived from precipitation reactions. The precipitating phase in Stainless W is believed to be titanium carbide and an aluminum-rich phase; in 17-4PH, the precipitating phase involves copper and may be similar to the epsilon phase of the iron-copper system.

The ingots, billets, and bars available for forging are generally processed using procedures similar to those for austenitic stainless steels. Since the alloys are not used extensively in the form of die forgings, mill suppliers generally stock a few standard sizes and either cog or hot roll them to the sizes specified by the forging companies.

With the exception of AM 355, the semiaustenitic, precipitation-hardenable stainless steels are produced by standard arc-melting practices. The AM 355 alloy usually is produced by consumable-electrode, vacuum-arc melting.

The martensitic precipitation-hardenable stainless steel forging billets usually are supplied in the overaged condition, which is developed by holding solution-treated material at 1150 F. Such material must be solution treated prior to precipitation hardening. Material supplied in the solution-treated condition may be precipitation hardened directly after forging.

FORGING BEHAVIOR

The composition of these alloys is such that some delta ferrite is usually formed during solidification of the ingot. During hot breakdown forging, the amount of delta ferrite may either decrease or increase, depending upon the temperatures employed. In the finished product, the semiaustenitic, precipitation-hardenable stainless steels may contain as much as 10 to 20 per cent delta ferrite. Under proper forging conditions, martensitic alloys will not contain delta ferrite.

The horizontal line on Figure 12-1 represents the start of carbide precipitation which usually commences at about 1750 F. Below this line, the forgeability of the

alloys drops sharply. This region is considered a lower limiting forging temperature. An upper limiting forging temperature occurs when additional delta ferrite forms on heating. Alloy compositions represented in Figure 13-1 by the right-hand vertical line exhibit poor forgeability because they retain some delta ferrite regardless of temperature. The delta ferrite, of course, is detrimental to the transverse ductility, particularly in the short transverse direction.

According to forging sources, the AM 355 alloy has better forgeability than the similar AM 350 alloy. The AM 355 alloy has a lower chromium and a higher carbon content than the AM 350 alloy. The AM 350 composition has a greater tendency to exhibit delta ferrite over the entire forging range as would be expected from the data in Figure 13-1.

The relative forging behavior of the precipitation-hardenable stainless steels can be illustrated by the following comparison of forging characteristics between AISI 4340 steel and the 17-7PH, AM 355 and 17-4PH steels:(5)

	4340	17-7PH	AM 355	17-4PH
Forging temperature	2300 F	2150 F	2150 F	2150 F
Decarburization	High	Low	Low	Low
Scale	High	Low	Low	Low
Grain-size control	Excellent	Fair	Fair	Good
Forgeability	Excellent	Fair	Good	Good
Forging pressure (relative)	1.0	1.4	1.4	1.4
Thermal cracking	Low	None	Low	Medium
Die wear	Low	Medium	Medium	Medium

Because of the combination of lower forging temperature and greater stiffness, 30 to 50 per cent higher forging loads are required for the precipitation-hardenable stainless steels than for 4340 and, accordingly, heavier equipment is needed. On the other hand, the precipitation-hardenable grades are much less sensitive to decarburization than are the higher carbon low-alloy steels. Also, they do not scale as much. Thus, it is possible to design some precipitation-hardenable stainless steel forgings for use with as-forged surfaces.

INFLUENCE OF FORGING VARIABLES ON MECHANICAL PROPERTIES

Forging procedures play an important role in achieving the optimum mechanical properties in precipitation-hardenable stainless steels, particularly in the semiaustenitic grades in which the grain size is primarily controlled by prior processing. For example, forging the AM 355 alloy at too high a finishing temperature produces a coarse grain structure. The coarse grains may not transform properly during the subsequent heat treatment, and an undesirable amount of austenite may be retained in the finished component.

The quality of semiaustenitic-alloy forgings is also quite sensitive to the amount of forging reduction as well as the forging temperature. For this reason, according to a forging producer, the mechanical properties of as-received billet stock do not give an accurate measure of material quality as it might later apply to forgings. In one

case cited, a heat of 17-7 PH was used to make two different forging configurations. Both forgings were dish-shaped, but one contained relatively heavy sections which received only minor reduction during forging. This latter component exhibited very erratic mechanical properties after the same thermal treatment as that given the other forging. In ultrasonic tests the heavy-section forging gave many indications of sonic defects but the well-worked forging was virtually clear of these defects. The ultrasonic defects and the lack of adequate response to heat treatment were attributed to a high percentage of retained austenite in forgings with heavier sections.

A similar comparison of forgings of AM 355 steel was reported by the same producer with respect to the effects of forging variables. It was found that the final forging temperature must be held low enough to insure complete transformation, and that when final forging temperatures were much above 2000 F considerable retention of austenite occurred.

In addition to improving forgeability, it is important to control the amount of delta ferrite present to retain satisfactory short transverse ductility levels. The typical effect of delta ferrite on the transverse tensile ductility of semiaustenitic alloys is illustrated by the following data:

Alloy	Ferrite Content, per cent	Ultimate Strength, 1000 psi	Elongation, per cent
AM 355	0	178	22.0
17-7PH	0	184	19.0
17-7PH	0 to 2	180	5.0 to 8.0
17-7PH	Over 2	180	0 to 3
PH15-7Mo	About 2	205	1 to 4
PH15-7Mo	About 2	195	4 to 9

Since only small amounts of delta ferrite in the structure have a serious effect on transverse ductility, close control of the forging temperature and strain rate for a given reduction are necessary to avoid overheating. Delta ferrite also can be formed by overheating the martensitic alloys, so this holds true for the precipitation-hardenable stainless steels in general.

COMMENTARY ON FORGING PRACTICES

The precipitation-hardenable stainless steels are not used as extensively in the form of forgings as are the martensitic low-alloy steels such as AISI 4340, D6AC, etc.(6) Consequently, fewer quantitative data are available on forgeability, forging pressure, or forging design.

Since the basic alloy contents of these steels are quite similar to those of the 18-8 variety of austenitic stainless steels, forging pressure requirements are quite similar. The two-phase nature of alloys like PH15-7Mo, 17-7PH, and AM 350 gives rise to variable forgeability response. At temperatures above about 1700 F the AM 355 grade exhibits forgeability comparable to the 18-8 stainless. Below 1700 F, however, the forgeability drops rapidly due to carbide precipitation, as pointed out in the discussion of Figure 12-1.

The forging characteristics of the martensitic grades are about the same as the 12-14 per cent chromium stainless steels (e.g., Type 410), according to a major forging producer. Care must be exercised in forging 17-4PH to avoid overheating at the center of heavy sections from too heavy or too rapid reductions, or thermal cracking can occur. Excellent reproducibility of both forgeability and mechanical properties, however, are attainable with 17-4PH.

The semiaustenitic precipitation-hardenable alloys frequently require additional forging steps to allow for conditioning to remove cracks. This is brought out in Table 12-2 which compares the operations for forging a typical part from 17-7PH, 17-4PH, and 4340 steels. Here it can be seen that the processing of the martensitic alloys and the low-alloy steels is quite similar.

TABLE 12-2. COMPARISON OF FORGING OPERATIONS FOR 17-7PH, 17-4PH, AND 4340 STEELS ON A TYPICAL PART

17-7PH	AISI 4340	17-4PH
Cut stock	Cut stock	Cut stock
Heat 2150 F	Heat 2300 F	Heat 2150 F
Upset one end	Upset one end	Upset one end
Cool to room temperature		
Remove ruptures		
Heat 2150 F	Reheat 2300 F	Reheat 2150 F
Flatten and draw one end	Flatten and draw one end	Flatten and draw one end
Cool to room temperature		
Inspect and remove ruptures		
Heat 2150 F	Reheat 2300 F	Reheat 2150 F
Block	Block	Block
Reheat 2150 F		Reheat 2150 F (optional)
Finish and trim	Finish and trim	Finish and trim
		Equalize 2150 F
Air cool	Slow cool	Slow cool

All of the semiaustenitic grades have a tendency to retain the austenite structure at room temperature after forging at temperature higher than 2000 F and heat treating. To achieve uniform grain refinement it is frequently necessary to promote the austenite transformation to martensite by heating forgings in the vicinity of 1400 F. This treatment causes carbide to precipitate, renders the austenite less stable, and promotes complete transformation to martensite on cooling back to room temperature. Forgings heat treated in this way are in a favorable condition for subsequent heat treatment and grain refinement.

The semiaustenitic grades are sensitive to composition variations. The PH15-7Mo alloy, for example, may exhibit delta ferrite contents ranging from 0 to 15 per cent at room temperature depending on composition. It is, therefore, common practice to determine the delta ferrite content before forging. If billets exhibit stringers of ferrite on their surfaces they will usually rupture during any sizeable forging reduction. Ferrite located in the center of the billets does not have as marked effect on forgeability, although it will reduce the transverse ductility of the forgings.

Selection of Forging Temperatures. When selecting forging temperatures for these alloys, it is important to consider such factors as the amount and rate of reduction, grain growth, the delta ferrite transformation, and the nature of subsequent heat treatments. For example, AM 355 alloys can be forged at temperatures as high as 2200 F without encountering ferrite cracking. However, temperature increases occurring during either rapid or large reductions can cause delta ferrite to form and, thus, promote cracking. The alloys generally exhibit rapid grain growth at temperatures above 2100 F. Forging temperatures for parts receiving small reductions are therefore maintained below 2100 F. This is particularly important for the alloys that retain austenite to room temperature because subsequent heat treatment will not provide grain refinement.

Forging temperatures recommended for several precipitation-hardenable stainless steels are given in Table 12-3. When forging rapidly with large reductions in a single operation it is good practice to lower the forging temperature 100 F or more to prevent the formation of additional delta ferrite resulting from temperature increases during forging. This would be particularly true if forging is done in high-energy-rate machines.

TABLE 12-3. FORGING TEMPERATURES RECOMMENDED FOR PRECIPITATION-HARDENABLE STAINLESS STEELS

Alloy	Maximum Forging Temperature, F	Recommended Temperatures for Forgings Receiving Given Nominal Amounts of Reduction, F			
		Light (Up to 15%)	Moderate (15-50%)	Severe (Over 50%)	Variable(a)
			Semiaustenitic		
AM 350	2150	2100	2150	2150	2100
AM 355	2200	2000	2150	2150	2000
17-7PH	2200	2050	2150	2200	1950
PH15-7Mo	2250	2000	2100	2150	2000
			Martensitic		
Stainless W	2250	2050	2200	2200	2050
17-4PH	2200	2100	2150	2150	2100

(a) Variable Reduction - This refers to forgings receiving widely differing reductions. End upsets, for example, receive large reductions on the upset end while the shaft may remain essentially undeformed.

Heating to the Forging Temperature. To avoid internal cracks, cold thick sections should not be charged directly into a furnace heated at the forging temperature, but should be preheated at 1200-1400 F and allowed to equalize at this temperature prior to charging into the hot furnace. For example, the extent of the preheating that is required to avoid high internal stresses is prescribed as follows for the 17-4PH grade:

Billet Thickness, inches	Maximum Furnace Temperature at Time of Charging, F
<4	2150
4-6	2000
6-[illegible]	1800
8-10	1600
10-12	1400
>12	1200

If reheating is necessary during forging, the steel should be charged into a furnace at 2150 F (the forging temperature) and held 1/2 hour for each inch of billet thickness to allow the center to reach temperature. Billets 3 inches and larger should soak 1 hour at temperature before forging. Smaller billets should be soaked for a minimum of 15 minutes. Because of the formation of delta ferrite, overheating should be avoided. Although these recommendations are for the 17-4PH grade, similar precautions should be followed for the other precipitation-hardenable stainless steels as well.

Cooling After Forging. Cooling after forging for the stainless steels should be controlled to minimize the chances for cracking. This is particularly important for the martensitic grades which transform on cooling. Sections thicker than three inches and intricate smaller sections should be returned to the heating furnace and equalized at the forging temperature prior to cooling. Sections smaller than 6 inches may then be air cooled. Larger sections should be cooled more slowly, and this can be simply accomplished by cooling under a cover of light-gage steel sheet.

Furnace Atmospheres. Because careful control of both carbon and nitrogen is imperative for these steels, heating in reducing atmospheres whereby either carbon or nitrogen may be picked up by the steel is considered poor practice. Heating in cracked ammonia is also not advisable. Furthermore, carbonaceous lubricants can give rise to carburization. For this reason, the forging billets must be thoroughly cleaned prior to heating since any carburization or nitrogen pickup will produce a more stable austenite and will significantly reduce the strength properties obtained on heat treatment.

Oxidizing atmospheres are recommended for all heating operations. This will result in scaling. The scale formed in an oxidizing atmosphere, however, is readily removed by pickling in a nitric-hydrofluoric acid mixture.

If the part is formed to finish shape, acid pickling is not recommended since this will result in some intergranular attack around the carbides at the delta ferrite-austenite interfaces as well as at the prior austenite grain boundaries. In this case, scale removal by vapor blasting is recommended.

Lubrication. Lubrication techniques for these alloys are essentially the same as those for the austenitic stainless steels. When forging parts containing thin webs, lubrication should be controlled to reduce surface metal flow. Any rupturing that might occur from the presence of delta ferrite would be exaggerated by large deformations of the surface.

Coining. The semiaustenitic stainless steels retain the soft austenite structure to room temperature after appropriate heat treatments. This permits a certain amount of cold forging or coining to close dimensional tolerances. Since the austenite is metastable, large cold reductions may lead to cracking from strain-induced transformation. Reductions in the neighborhood of 20 per cent are considered safe for coining the semiaustenitic steels, although no quantitative data are available to confirm this. During the hardening treatment, however, a growth of about 0.004 inch per inch occurs. In cold coining to final dimensions, this growth factor must be taken into account in the

process design. The martensitic precipitation-hardenable stainless grades are not amenable to cold coining.

REFERENCES

(1) Ludwigson, D. C., and Hall, A. M., "The Physical Metallurgy of Precipitation Hardenable Stainless Steel", DMIC Report No. 111, Defense Metals Information Center, Battelle Memorial Institute (April 20, 1959).

(2) Ludwigson, D. C., "Semiaustenitic Precipitation-Hardenable Stainless Steels", DMIC Report No. 164, Defense Metals Information Center, Battelle Memorial Institute (December 6, 1961).

(3) Smith, R., Wyche, E. H., and Gorr, W. W., "A Precipitation-Hardening Stainless Steel of the 18 Per Cent Chromium, 8 Per Cent Nickel Type", Tech. Publ. No. 2006, Metals Technology, 13 (June 1946).

(4) Lena, A. J., "Precipitation from Solid Solution", Precipitation Reactions in Iron-Base Alloys, American Society for Metals, Cleveland, Ohio (1959).

(5) Henning, H. J., and Boulger, F. W., "High-Strength Steel Forgings", DMIC Report No. 143, Defense Metals Information Center, Battelle Memorial Institute (January 5, 1961).

(6) Henning, H. J., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", Technical Documentary Report No. ML-TDR-64-95, Contract No. AF 33(600)-42963, Battelle Memorial Institute (March 1964).

CHAPTER 13

MARAGING STEELS

Crystalline Structure:	Austenite – Face-centered cubic Martensite – Body-centered cubic*
Phase Changes:	Austenite stable to about 300 F, below which it transforms to martensite
Number of Phases:	One phase (austenite) at forging temperatures
Solidus Temperature Range:	Above 2500 F
Reactions When Heated in Air:	Oxidation

ALLOYS AND FORMS AVAILABLE

Since the maraging steels were first announced in 1959, six wrought steels and one cast steel of this type have been developed.(1) The compositions of these alloys are shown in Table 13-1.(2, 3, 4) As their name implies, the maraging steels develop their ultrahigh strength by the combination of transformation to martensite followed by age hardening.(1, 5) The 25 per cent nickel steel is austenitic at room temperature after the standard anneal at 1500 F. To martensitize this steel, it is reheated to 1100 to 1300 F which depletes the matrix of nickel by promoting the precipitation of nickel-containing intermetallic compounds. The austenite thus depleted is sufficiently unstable for considerable martensite transformation to occur on cooling to room temperature.(1) Cooling to -100 F is required for complete transformation to martensite. The 20 per cent, 18 per cent, and 15 per cent nickel types (the last named being developed for elevated-temperature applications up to 1000 F) are martensitic as annealed or as hot worked. The standard annealing temperature for these steels is 1500 F. For complete martensitization, the 20 per cent nickel steel is refrigerated at -100 F; transformation in the 18 per cent and 15 per cent nickel steels is substantially complete at room temperature. All of the maraging steels undergo tremendous increases in strength on aging in the range of 750 to 1000 F after martensitization. An aging treatment commonly used for the 18 per cent nickel type is 900 F for 3 hours.(6)

The 18 per cent nickel type of maraging steel has received the greatest attention because of its outstanding fracture toughness and because it is hardened by a simple single-step schedule carried out at a moderate temperature. In addition, because of their low carbon content, the maraging steels as a class are not troubled with decarburization problems; they are readily fabricated and reasonably weldable; they are not

*The martensites usually formed in steels are tetragonal iron-carbon martensites; the martensites developed in the maraging steels are essentially carbon-free, iron-nickel martensites which are body-centered cubic.

sensitive to heating or cooling rate either in annealing or in aging; and they are subject only to minor dimensional changes on hardening.

TABLE 13-1. COMPOSITIONS OF NICKEL MARAGING STEELS

Alloy Designation	Composition, per cent[a]										
	C[b]	Mn[b]	Si[b]	S[b]	P[b]	Ni	Co	Mo	Ti	Al	Cb
25 Ni	0.03	0.10	0.10	0.01	0.01	25.0-26.0	--	--	1.3-1.6	0.15-0.30	0.30-0.50
20 Ni	0.03	0.10	0.10	0.01	0.01	19.0-20.0	--	--	1.3-1.6	0.15-0.30	0.30-0.50
18 Ni (280)	0.03	0.10	0.10	0.01	0.01	18.0-19.0	8.5-9.5	4.6-5.2	0.5-0.8	0.05-0.15	--
18 Ni (250)	0.03	0.10	0.10	0.01	0.01	17.0-19.0	7.0-8.5	4.6-5.2	0.3-0.5	0.05-0.15	--
18 Ni (200)	0.03	0.10	0.10	0.01	0.01	17.0-19.0	8.0-9.0	3.0-3.5	0.15-0.25	0.05-0.15	--
Cast	0.03	0.10	0.10	0.01	0.01	16.0-17.5	9.5-11.0	4.4-4.8	0.15-0.45	0.05-0.15	--
15 Ni[c]	0.01	--	--	--	--	15	9	5	0.7	0.7	--

(a) Other elements added: 0.003 B, 0.02 Zr, and 0.05 Ca.
(b) Maximum.
(c) Nominal composition.

Thus far, three grades of 18 per cent nickel maraging steel have been developed. Their nominal room temperature yield strengths as annealed and fully aged are 280, 250 and 200 ksi and the corresponding designations for them are 18 Ni (280), 18 Ni (250), and 18 Ni (200). The 18 Ni (280) steel is consumable-electrode vacuum-arc remelted because of its high titanium content. The other grades are air melted, air melted and vacuum degassed, or consumable-electrode vacuum-arc remelted. Ingots, billets, and bars are generally processed by procedures similar to those used for alloy steels containing substantial amounts of nickel, i.e., the 3 to 9 per cent nickel steels. Available forms include forging billets in sizes to 12 inches diameter and hot rolled bar stock in a wide range of sizes in addition to plate, sheet, strip, seamless pipe and wire.

FORGING BEHAVIOR

The 18 Ni maraging steels are readily hot worked by conventional methods and standard equipment. These steels are heated at 2200 to 2300 F, and forging operations may be started at temperatures as high as 2300 F.[7] It is generally recommended that the finishing temperature be in the vicinity of 1700 to 1500 F. The materials do not exhibit hot shortness up to 2300 F and do not crack even when finished below 1500 F.[7]

The martensite formed in the 18 Ni maraging steels is relatively soft and ductile before being aged. For bar stock, the yield strength is in the order of 100,000 psi and the elongation in 1 inch is about 17 per cent.[7] Likewise, the tendency for this material to work harden is extremely low. Therefore, because of these factors, the 18 Ni maraging steels should be readily amenable to cold forging operations such as coining.

INFLUENCE OF FORGING VARIABLES ON MECHANICAL PROPERTIES

The effect which hot forging has on the microstructure and mechanical properties of the 18 Ni maraging steels depends on (a) the type of forging being done, (b) the forging temperature, and (c) the extent to which the metal is worked. The effects are similar to those produced in the austenitic-stainless steels and other malleable single-phase metals and alloys.

The influence of hot working (swaging) on the grain size of 18 Ni (280) maraging steel is illustrated by the data in Table 13-2.[8] The data indicate that, between 1500 F and 2200 F, the effect of temperature is almost negligible, but the influence of per cent reduction is quite considerable. It should be noted that these alloys do not recrystallize when austenitized i the way that carbon and low-alloy steels do; the grain size after annealing is generally the same as after forging. The principal method available to control grain size is through adjustment of the hot-working conditions. In addition, it is possible that sufficient residual strain can be developed in the material by heavy working at the low-temperature end of the forging range so that, on reheating to 1500 to 1700 F, recrystallization will occur.

TABLE 13-2. INFLUENCE OF SWAGING TEMPERATURE AND REDUCTION ON GRAIN SIZE OF 18 Ni (280) MARAGING STEEL

Swaging Temperature, F	ASTM Grain Size[a] After Indicated Swaging Reduction and Anneal[b] 35%	74%
1500	2-4	9-10
1600	1-2	10
1700	2-3	10
1800	3-6	9-10
1900	4-6	7-10
2000	2-4	8-10
2100	2-4	7-10
2200	2-3	7-9

(a) Grain size observed on surface normal to axis of bar.
(b) Bars annealed at 1500 F and aged at 900 F.

The tensile properties obtainable in closed-die press forgings made from 18 Ni maraging steels are illustrated by the data in Tables 13-3 and 13-4.[9] The data in Table 13-3 relate to a 40-inch-diameter front closure for the Pershing rocket motor case which was forged from a consumable-electrode vacuum-arc remelted heat of the 18 Ni (280) grade of maraging steel. The test specimens were taken from the rim section. It is seen that while the strength values are about the same in both directions, the ductility in the radial direction tends to be less than in the tangential direction.

TABLE 13-3. TENSILE PROPERTIES OF 18 Ni (250) MARAGING STEEL PRESS-FORGED DOME AS ANNEALED AT 1500 F AND AGED 3 HOURS AT 900 F AFTER FORGING[9]

Direction[a]	0.2% Offset Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation, per cent	Reduction of Area, per cent
Tangential	284.5	294.7	9.5	44.7
"	288.0	297.6	10.0	45.3
Radial	285.6	292.7	8.0	35.4
"	281.5	290.7	8.0	42.3

(a) Specimens removed from the region of the rim of the dome which was 40 inches in diameter.

TABLE 13-4. TENSILE PROPERTIES OF A FORGED 18 Ni (250) MARAGING STEEL AIRCRAFT STRUCTURAL COMPONENT AS ANNEALED AT 1500 F AND AGED 3 HOURS AT 900 F[9]

Location	Direction	0.2% Offset Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation, per cent	Reduction of Area, per cent
Center heavy section	Short transverse	250.9	263.1	5.0	26.3
Ditto	Short transverse	252.7	264.9	4.5	21.5
"	Longitudinal	250.9	263.1	10.0	46.7
"	Longitudinal	250.9	263.9	10.0	48.2
"	Transverse	247.8	260.1	5.0	18.8
"	Transverse	251.9	263.5	4.0	9.6
Web area	Longitudinal	252.9	265.2	10.0	52.9
Ditto	Longitudinal	254.5	267.2	11.0	53.5
"	Transverse	248.8	260.8	10.0	47.5
"	Transverse	252.5	264.3	10.0	48.7

Table 13-4 presents tensile data for an aircraft structural component press forged from a consumable-electrode vacuum-arc remelted heat of the 18 Ni (250) grade. The strength values are remarkably similar for all three directions and for both locations sampled. However, the ductility falls off significantly in the transverse and short transverse directions, except that the transverse ductility in the thin webbed area is about the same as the longitudinal ductility. The low ductility in the transverse and short transverse directions of the center section is thought to be associated with the fact that very little reduction occurred in these directions in the center of the part during the final

forging operation.[9] The web area, on the other hand, received a considerable amount of deformation during the finish forging operation. These variations can be corrected to a large extent by tailoring the forging process in all its details to the material being forged.[9]

Data illustrating the influence of forging temperature on the tensile strength of the 18 Ni type of maraging steel are shown in Figure 13-1.[8] It is to be noted that the strength of material aged directly after forging increases significantly for forging temperatures below 1700 F; whereas, when the steel is annealed before aging, the strength is essentially independent of forging temperature. As a result, for material forged at temperatures above 1700 F, annealing at 1500 F before aging improves properties to some extent; on the other hand, for material forged at temperatures below 1700 F, this annealing treatment reduces strength. Quite probably, forging at temperatures below 1700 F imparts considerable strain hardening to the metal, which accounts for its greater strength when forged at such temperatures. A thorough annealing treatment would be expected to remove this strain-hardening effect and, in this way, cause a reduction in strength.

From the data in Figure 13-1, it would seem that this type of steel should be finish forged at or below 1600 F and aged directly, in order to develop maximum strength. However, in complicated forgings, the degree of strain hardening is apt to vary greatly from section to section and the mechanical properties are likely to vary accordingly. The data suggest that much more uniformity of strength properties throughout a forged component is to be expected if the metal is annealed before aging.

COMMENTARY ON FORGING PRACTICE

The selection of forging temperatures for the maraging steels is dominated by the type of forging operation intended. If the part is designed in such a way that it will receive sizeable reductions, forging temperatures are similar to those used for alloy steels. If the part contains portions that receive little reduction during forging, however, temperatures should be adjusted downward to below 2000 F. This is done to prevent excessive grain growth. In any case, the final forging steps should be completed below 2000 F. Recommended forging temperatures are given below for some typical forgings made from the 18 Ni alloys.

Type of Forging	Recommended Forging Temperatures
Disks	Upset 2300 F; Finish 2150 F
Rib and web shapes	Draw 2200 F; Block 2250 F; Finish 2150 F
End upset	Upset one end 2000 F
Rings	Upset and pierce 2200 F, Roll 2200 F

The end upset generally receives deformation only on one end. Thus, the lowest forging temperature is recommended for such parts.

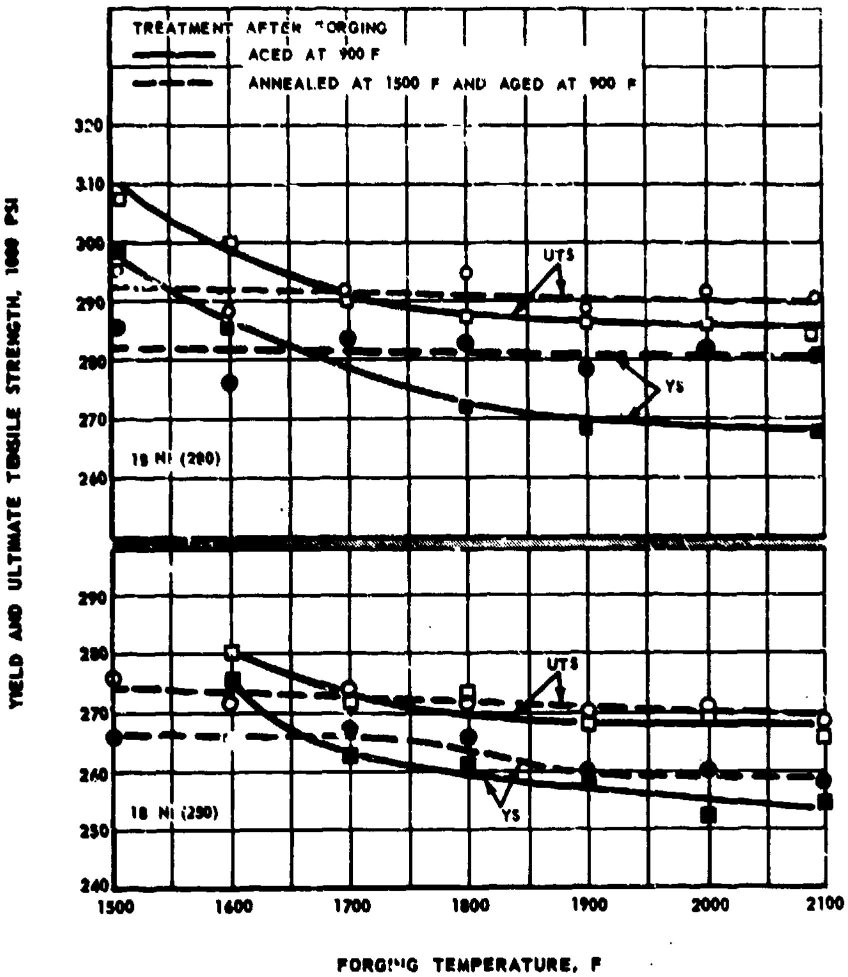


FIGURE 13-1. EFFECT OF FORGING TEMPERATURE AND SUBSEQUENT HEAT TREATMENT ON STRENGTH OF 18 Ni MARAGING STEELS

After DeVries(8)

Because lower finishing temperatures (1500 to 1700 F) are often needed for grain-size control and ductility, higher forging pressures are required than normal for other steels. Sparks reports that forging loads are comparable to some titanium alloys.[10] Otherwise, forging procedures normally used for alloy steels are suitable for the 18 Ni maraging steels.

REFERENCES

(1) Bieber, C. G., "Progress with 25 Per Cent Nickel Steels for High Strength Applications", Metal Progress, 78, 99-100 (November, 1960).

(2) Drennen, D. C., and Roach, D. B., "Properties of Maraging Steels", Defense Metals Information Center, Battelle Memorial Institute, DMIC Memorandum 15((July 2, 1962).

(3) Sadowski, E. P., and Decker, R. F., "Cast Maraging Steel", Modern Castings, 43, pp 26-35 (February, 1963).

(4) Floreen, S., and Decker, R. F., "Maraging Steel for 1000°F Service", Trans. ASM, 56, pp 403-411 (September, 1963).

(5) Decker, R. F., Eash, J. T., and Goldman, A. J., "Eighteen Per Cent Nickel Maraging Steel", Trans. ASM, 55, 58-76 (March, 1962).

(6) Floreen, S., and Decker, R. F., "Heat Treatment of 18 Ni Maraging Steel", Trans. ASM, 55, pp 518-530 (September, 1962).

(7) The International Nickel Company, Inc., "Interim Data Sheet on 18 Per Cent Nickel Maraging Steel", 20 pp (November 26, 1962).

(8) De Vries, R. P., "The Effect of Processing Factors and Size on the Properties of Commercially Produced 18 NiCoMo Maraging Steels", Symposium on Maraging Steels, Wright-Patterson Air Force Base, Ohio (May 14, 1962).

(9) Sparks, R. B., "Properties of Maraging Steel Press Forgings", Maraging Steel Project Review, ASD Technical Documentary Report 63-262, 196-222 (May, 1963).

(10) Sparks, R. B., "Working With Maraging Steels . . . . Forging", Metal Progress, 83, No. 7 (July, 1963).

CHAPTER 14

TITANIUM AND TITANIUM ALLOYS

Crystalline Structure:	Alpha – Close-packed hexagonal Beta – Body-centered cubic
Phase Changes:	On cooling from solidus region, beta phase transforms partly or wholly to alpha
Number of Phases Present:	Two; predominantly beta at forging temperatures; alpha generally present in amounts varying up to about 50 per cent
Liquidus Temperature Range:	3000 to 3135 F
Solidus Temperature Range:	2700 to 3100 F
Reactions When Heated in Air:	Continuous oxidation and gas absorption above about 1200 F

ALLOYS AND FORMS AVAILABLE

There are three general types of titanium-base alloys: (1) alpha alloys, (2) alpha-beta alloys, and (3) beta alloys. Forging behavior and, hence, forging practice are controlled by the alpha-beta relationship of each alloy type. On heating, pure titanium undergoes a transformation at 1625 F from a hexagonal close-packed structure (alpha)

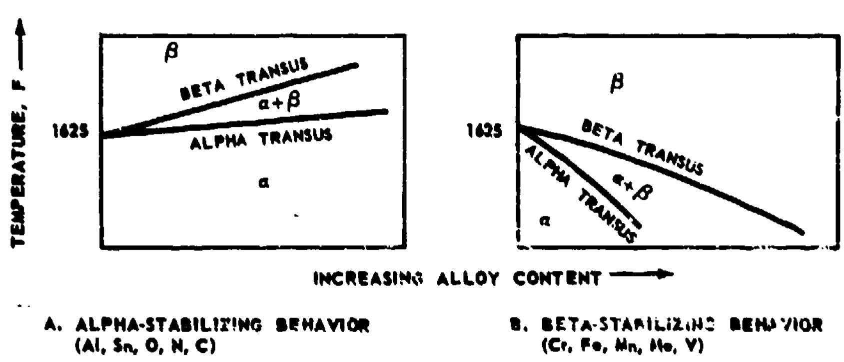


FIGURE 14-1. ALPHA-BETA RELATIONSHIPS FOR THE TWO PRINCIPAL ALLOYING BEHAVIORS OF TITANIUM

to a body-centered cubic structure (beta). Alloying elements are added to stabilize one or the other of these phases as shown in Figure 14-1, by either raising or lowering the transformation temperatures.

Aluminum, which is the strongest metallic alpha stabilizer used in commercial alloys, raises the beta transus and imparts high-temperature strength to the alloy. Certain alloying elements used in commercial alloys (e.g., Cr, Fe, Mn, Mo, V) are beta stabilizers which lower the beta transus. Additions of these elements strengthen the beta solid solution and increase the amount of beta retained at room temperature. Thus, the beta may be changed from an unstable form to a stable form, even below room temperature. In alpha-beta alloys the beta phase is in an unstable condition. These alloys can be heat treated to achieve strengthening by partial transformation of the beta phase to alpha, which is finely dispersed in the beta phase. The beta transus generally establishes the maximum temperature to which these alloys can be safely heated for optimum mechanical properties. Hence, most high-strength forging alloys are designed with sufficient amounts of beta stabilizers to retain beta in an unstable, heat-treatable form, and with additions of aluminum to increase the beta transus and thereby extend the forging-temperature range.

The compositions and individual forging characteristics of several titanium forging alloys are presented in Table 14-1. The alloys are produced commercially as cylindrical ingots by vacuum-arc melting in consumable-electrode furnaces. Ingots are first cogged to large, square billets, then hot rolled to bar of various sizes for forging.

FORGING BEHAVIOR

Since virtually all of the titanium forging alloys are double-melted, they rarely contain segregations of other materials that might cause wide variations in forgeability. Initial breakdown forging of titanium alloys is usually done at temperatures above the beta transus because the body-centered cubic structure is more ductile and forging-pressure requirements are generally lower. Final forging is usually done at temperatures below the beta transus to prevent excessive beta grain growth and attendant low ductility.

Variations in strain rate have little influence on the forgeability* of alpha and alpha-beta alloys; both alloy types are readily forgeable in either presses or hammers. The Ti-13V-11Cr-3Al beta alloy also exhibits good forgeability in both presses and hammers when forged above 1400 F. However, when forged just below 1400 F, the alpha phase begins to precipitate and the alloy is more susceptible to cracking, particularly in drop hammers. For reasons of mechanical-property control, this alloy is frequently forged in the lower temperature range but the reductions are small; hence, the reduced forgeability does not present a major problem.

The forging pressure of titanium alloys increases rapidly with increasing strain rate at forging temperatures. This effect is illustrated by the forging-pressure curves shown in Figure 14-2 for two different alloy types – alpha and beta – upset forged at different temperatures over a tenfold increase in strain rate.[1] At about 10 per cent

*The use of the term 'forgeability' in this text refers to the ability of a metal to be deformed without failure, regardless of the forging pressure.

TABLE 14-1. COMPOSITIONS AND FORGING CHARACTERISTICS OF TITANIUM AND SEVERAL TITANIUM ALLOYS

Alloy Designation	Unalloyed	Ti-5Al-2.5Sn	Ti-8Al-1Mo-1V	Ti-6Al-4V	Ti-7Al-4Mo	Ti-13V-11Cr-3Al
Forging Specifications	AMS 4921	AMS 4966	--	AMS 4928	--	--
Nominal Composition (Balance Titanium), %	99Ti	5Al 2.5Sn	8Al 1Mo 1V	6Al 4V	7Al 4Mo	13V 11Cr 3Al
Alloy Type	Alpha	Alpha	Alpha	Alpha-beta	Alpha-beta	Beta
Density, lb/cm^3	0.163	0.161	0.160	0.162	0.162	0.175
Solidus, F	3020	2820	2800	2800	2800	2700
Beta Transus, F	~1650	~1900	~1900	~1810	~1825	--
Recrystallization Temp, F	1200	1300	1300	--	--	1350
Forging Temp Range, F						
Initial Breakdown	1700-2000	1700-2000	1700-2000	1800-2000	1800-2000	1600-1900
Die Forging	1600-1700	1650-1850	1750-1950	1550-1800	1680-1850	1450-1700
Forgeability						
Presses	Good	Fair	Fair	Good	Good	Good
Hammers	Good	Fair	Fair	Good	Good	Fair
High-Energy-Rate Machines	--	--	--	Good	--	Fair

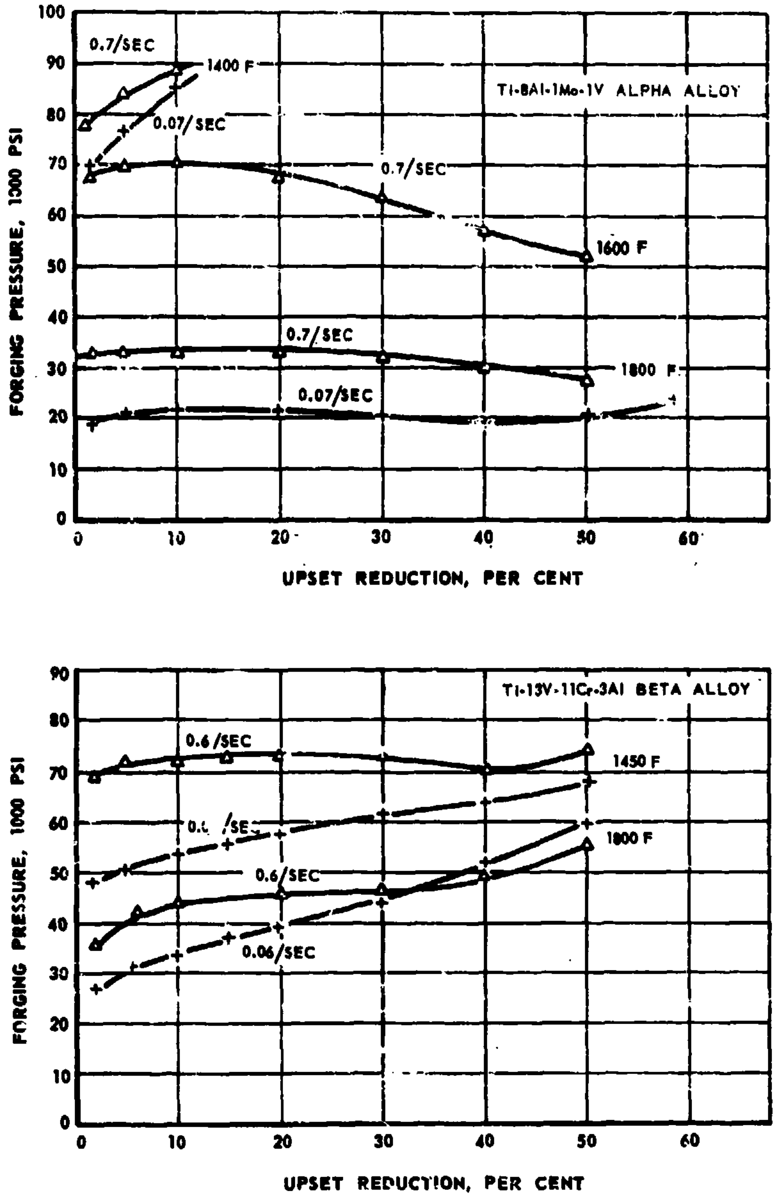


FIGURE 14-2. FORGING-PRESSURE CURVES FOR AN ALPHA AND A BETA ALLOY UPSET AT VARIOUS TEMPERATURES AND STRAIN RATES[1]

reduction, for example, the forging pressures for both alloys increase nearly 50 per cent for this moderate increase in strain rate.

Figure 14-3 summarizes the influence of strain rate and temperature on the forging pressure of several titanium alloys at approximate forging temperatures. Data are provided on a low-alloy steel (AISI 4340) for comparison. Forging pressure increases in an approximate linear relationship with the logarithm of the strain rate, as exemplified by the behavior of the Ti-13V-11Cr-3Al alloy. Thus, it is possible to estimate forging pressures for a given strain rate by interpolating or extrapolating data available at other strain rates.

An example of such an extrapolation is given in Figure 14-4 to permit a comparison of the forging pressures for Ti-6Al-4V alloy and AISI 4340 steel at their respective forging-temperature ranges. This indicates that Ti-6Al-4V at 1725 F requires about the same forging pressure at comparable deformation rates as AISI 4340 between 2000 F and 2300 F. At 1600 F, however, the data indicate that Ti-6Al-4V requires about twice the forging pressures of AISI 4340 steel. This coincides quite well with actual forging experience. Most forging companies report that Ti-6Al-4V requires one and one-half to two times the equipment capacity needed for forging alloy steels in comparable shapes.

The marked effect of temperature on the forging pressure of Ti-6Al-4V is characteristic of titanium alloys in general. Figure 14-5 shows how the forging pressure for each of the three types of titanium alloys increases more rapidly with decreasing temperature than does that of low-alloy steels. Thus, in ordinary die-forging operations, cooling of the workpiece has a more critical effect on forging pressures for titanium than for steels.

Since titanium alloys exhibit rapidly increasing strengths with increasing strain rate, more energy is required for hammer forging than for press forging at comparable temperatures. The Ti-13V-11Cr-3Al alloy at 1450 F, for example, requires nearly 50 per cent more energy at a typical hammer velocity of 200 in./sec than at a typical press velocity of 1.5 in./sec.(1) The energy required for forging Ti-13V-11Cr-3Al at hammer velocities at 1450 F is nearly four times that for AISI 4340 at 2300 F at the same level of reduction.

INFLUENCE OF FORGING VARIABLES ON MECHANICAL PROPERTIES

The mechanical properties of titanium alloys are basically a function of alloy content and thermal treatment. However, such properties as ductility, impact strength, and notch strength are influenced significantly by the forging procedures. Very important in the control of these properties is the control of microstructure which, in turn, is controlled by forging technique.

Alpha-Beta Alloys. The general influence of forging temperature on the room-temperature mechanical properties of alpha-beta alloys is illustrated in Figure 14-6. When forged above the beta transus, the alloys exhibit coarse beta grains with attendant low ductility and widely varying strength properties. Improved ductility and more uniform strength properties are obtained when the alloys are forged below the beta transus.

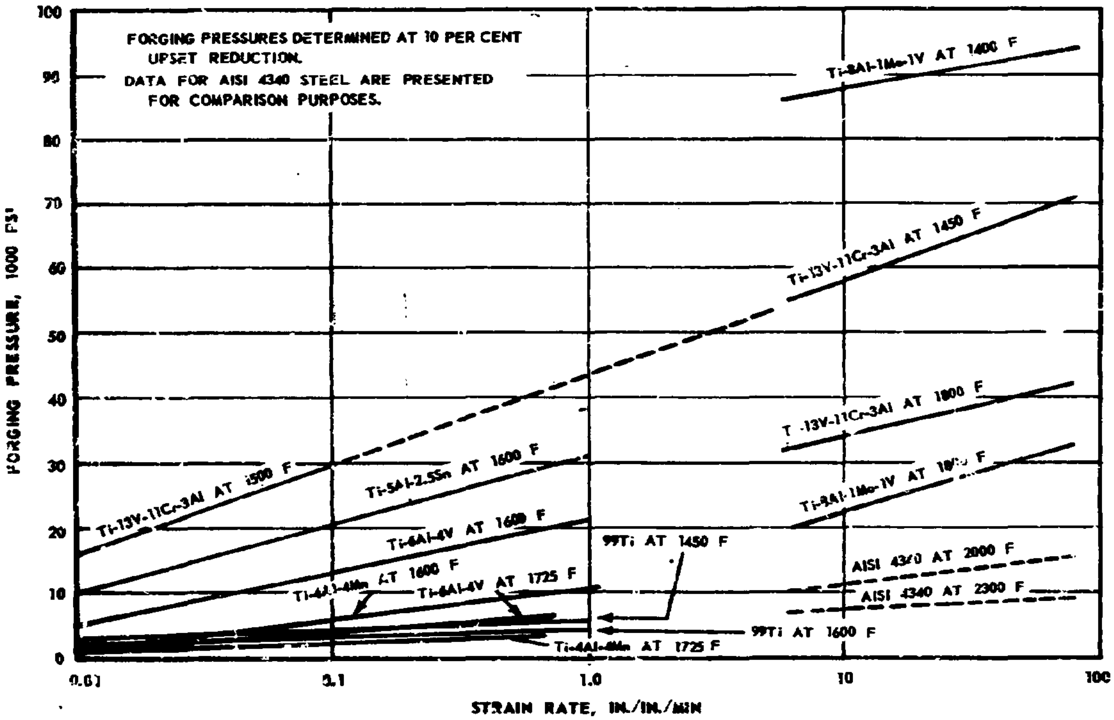


FIGURE 14-3. EFFECT OF STRAIN RATE ON FORGING PRESSURES FOR SEVERAL TITANIUM ALLOYS AT VARIOUS FORGING TEMPERATURES[1,2,3]

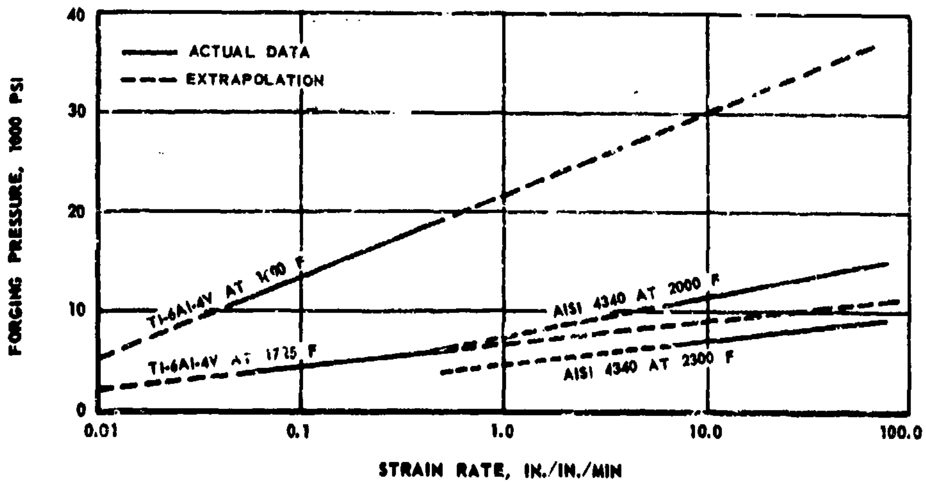


FIGURE 14-4. COMPARISON OF FORGING PRESSURES FOR TWO DIFFERENT MATERIALS BY EXTRAPOLATION OF FORGING-PRESSURE CURVES(1)

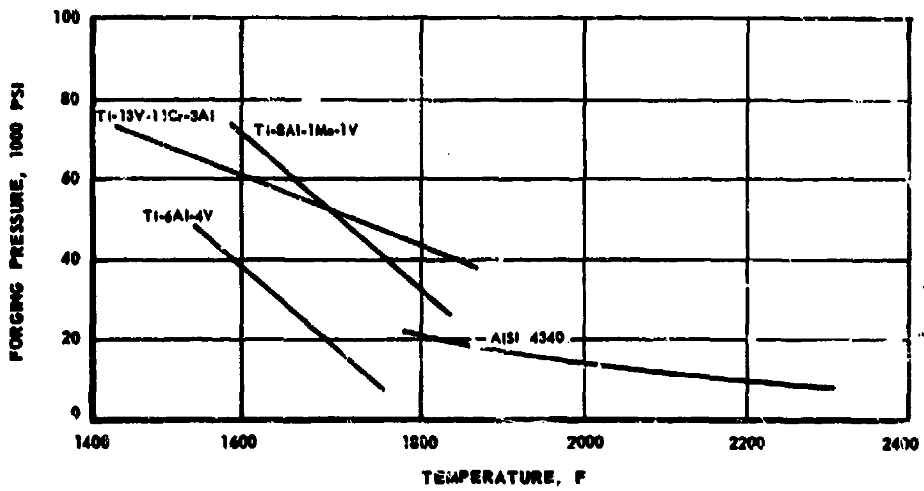


FIGURE 14-5. COMPARATIVE INFLUENCE OF TEMPERATURE ON THE FORGING PRESSURE OF TITANIUM ALLOYS AND LOW-ALLOY STEELS(1)

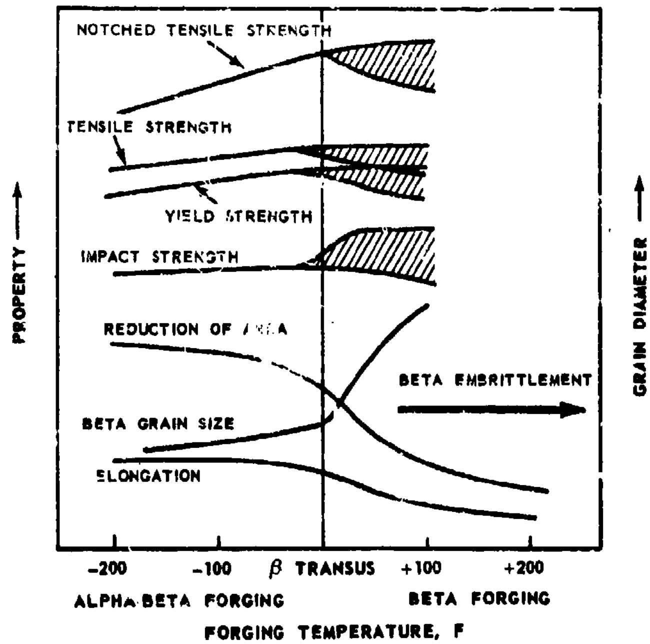
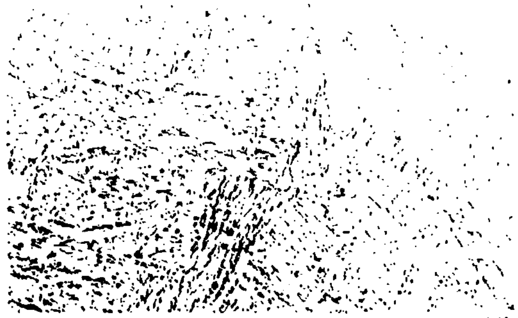


FIGURE 14-6. GENERALIZED INFLUENCE OF FORGING TEMPERATURE ON BETA GRAIN SIZE AND ROOM-TEMPERATURE MECHANICAL PROPERTIES OF FORGED ALPHA-BETA ALLOYS

Shaded areas indicate erratic behavior of beta-forged alloys.

This behavior is related to the physical characteristics of two-phase alloys. On heating above the beta transus, alpha-beta alloys transform completely to the beta phase. On cooling from the beta region, the beta decays or transforms partially to alpha, which takes on the classical acicular shape normal to such systems. If the beta-to-alpha transformation occurs during deformation, the alpha separating from the beta takes on an equiaxed grain structure (primary alpha). This is because the alpha repeatedly recrystallizes and undergoes grain growth at the expense of the transforming beta. However, if deformation is completed before the beta transforms to an equilibrium amount of alpha, the remaining beta transforms to the acicular alpha structure. Figure 14-7 illustrates the various typical structures obtained by forging at temperatures above and below the beta transus.

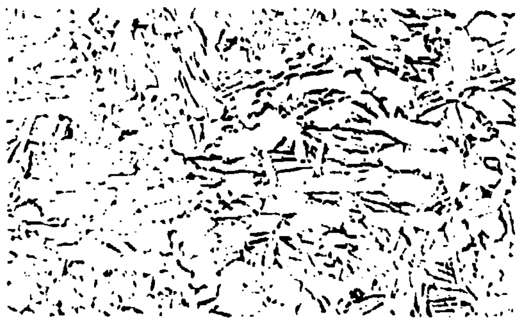
The relative amount of equiaxed alpha in the microstructure has an important influence on the ductility and notch sensitivity of alpha-beta alloys. Reduction of area increases with increasing amounts of equiaxed alpha up to about 20 per cent, then levels off; notch sensitivity steadily increases with increasing amounts of equiaxed alpha. These effects are illustrated in Figure 14-8 for the Ti-6Al-4V alloy. The best combination of properties is obtained when the structure contains about 20 to 30 per cent equiaxed alpha (Figure 14-7b). To achieve this structure, it is necessary to forge at temperatures about 100 F below the beta transus.



500X N32240

a. Forging Completed Above Beta Transus

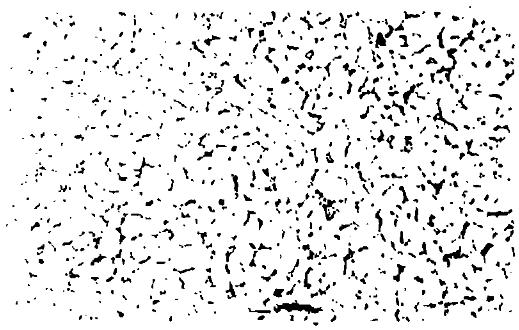
Structure: Acicular Alpha



500X N41095

b. Forging Completed About 100 F Below Beta Transus

Structure: About 30% Equiaxed Alpha, 50% Acicular Alpha, 20% Beta



500X N3224[illegible]

c. Forging Completed About 200 F Below Beta Transus

Structure: About 80% Equiaxed Alpha and 20% Beta

FIGURE 14-7. TYPICAL MICROSTRUCTURES OF ALPHA-BETA ALLOYS FORGED AT TEMPERATURES ABOVE AND BELOW THE BETA TRANSUS

Structures shown are for the Ti-6Al-4V alloy.

Thus, with material that has a prior beta structure it is important that a sufficient amount of work be done 100 F below the beta transus to obtain the desired microstructure. Colton determined that at least a 50 per cent reduction at this temperature is required to restore optimum room-temperature ductility in beta embrittled Ti-6Al-6V-2Sn alloy.(5) Hamer demonstrated a similar behavior for the Ti-7Al-4Mo alloy.(6) Figure 14-9 summarizes these findings and shows how elongation and reduction in area for these two alloys increase with increasing forging reductions.

These studies demonstrate that alpha-beta titanium alloys can be forged initially at temperatures above the beta transus, providing they receive at least 50 per cent final reduction at temperatures in the alpha-beta field.

Because of the critical influence of forging temperature on mechanical properties, it is important that the gross temperature changes that can be caused by the forging procedure are avoided. Large, rapid reductions, for example, can raise the actual forging temperature of a billet heated in the alpha-beta region to temperatures well above the beta transus and cause beta embrittlement. On the other hand, die-chilling effects can produce localized areas containing excessive amounts of the equiaxed-alpha phase. Hence, it is possible for a part forged at an "ideal" forging temperature to contain regions of varying structure from completely acicular to predominantly equiaxed alpha. Accordingly, the properties of such a forging will vary from an embrittled condition to a highly ductile, low-strength condition.

The structure, and hence properties, can also be affected by different cooling rates after forging. When forgings are water quenched immediately after forging, the beta phase transforms to a finer acicular structure. Slow cooling produces a structure containing grain boundaries rich in the alpha phase. Croan and Rizzitano found that Ti-6Al-4V alloy parts forged in the beta region exhibited higher ductility when water quenched than when air cooled.(7) Water quenching produced a fine acicular structure and, at the same time, appeared to arrest beta grain growth. This experience agrees well with the studies of Griest and Frost, which showed that the room-temperature tensile ductility of alpha-beta alloys decreases rapidly with increasing beta grain size.(8) The greatest loss of ductility was associated with continuous alpha at the grain boundaries.

Both investigators noted that notched-bar impact properties were virtually unaffected by beta-grain-size changes. In fact, the average impact properties at -40 F of Ti-6Al-4V alloy forgings containing an acicular structure were as much as 25 per cent higher than forgings containing a fine, equiaxed-alpha structure. Thus, for highest impact properties, it may be desirable to forge at temperatures above the beta transus and water quench immediately after forging. Such a practice, however, would be at the expense of other properties that suffer from beta forging.

If an unfavorably high proportion of equiaxed alpha is obtained by forging at low temperatures, it is possible to restore the desired proportion (2? to 30 per cent) by reheating to about 100 F below the beta transus and either air cooling or water quenching. This technique is useful for improving notch sensitivity at little or no expense of room-temperature ductility.

In summary, there are several alternatives available in the selection of forging temperatures for alpha-beta alloys for achieving optimum mechanical properties. These depend on the prior structure of the billet and the amount of reduction the various

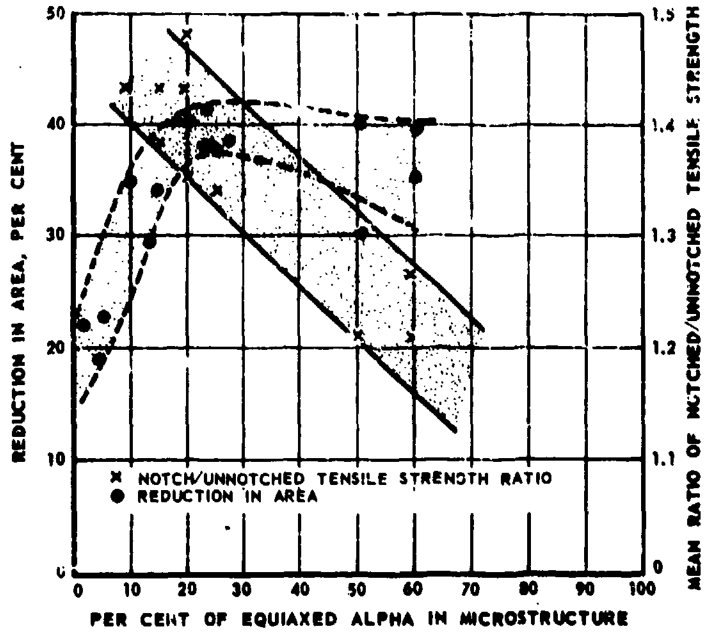


FIGURE 14-8. HOW DUCTILITY AND NOTCH SENSITIVITY OF Ti-6Al-4V ALLOY FORGINGS ARE INFLUENCED BY THE PROPORTION OF EQUIAXED ALPHA IN THE MICROSTRUCTURE[4]

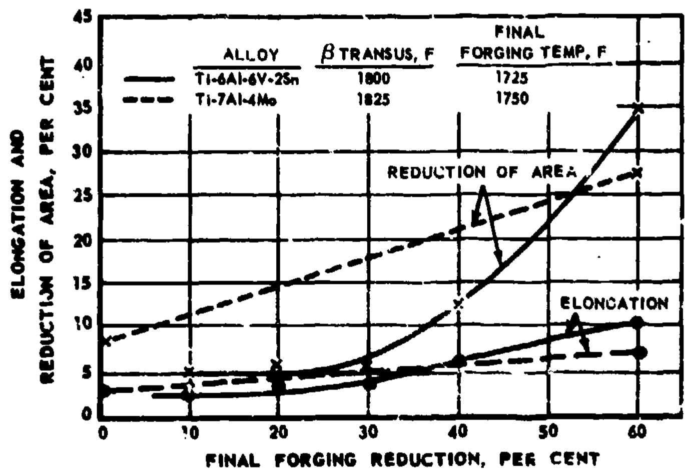


FIGURE 14-9. INFLUENCE OF FINAL FORGING REDUCTION BELOW THE BETA TRANSUS ON THE DUCTILITY OF TWO ALPHA-BETA ALLOYS HAVING A PRIOR BETA STRUCTURE[5,6]

parts of the forging will receive. Data and comments are presented in Table 14-2 for selecting the forging temperatures for two types of forgings — a disk shape and a rib-and-web structural part. As indicated, it is sometimes desirable to conduct preliminary open-die forging operations at temperatures in the beta region. This permits forging for a longer period of time and reduces the need for reheating. Beta forging would be recommended only if adequate reduction below the beta transus is possible in the subsequent forging operations. Temperatures for blocking operations are often adjusted downward to compensate for possible harmful temperature increases.

TABLE 14-2. CHOICE OF FORGING TEMPERATURES FOR TWO TYPICAL CONFIGURATIONS FORGED FROM ALPHA-BETA ALLOYS

Parts forged in a drop hammer
Beta transus of alloy: 1800 F

Operation	Nominal Linear Reduction, per cent	Recommended Forging Temperature(a), F	Remarks
			Disk Shape
Upset	70	1700	Temperature may increase as much as 150 F unless forging is done with light intermittent blows
Block	50	1650	More uniform reduction may be obtained by turning blank over between operations; should provide deformation in areas that exceeded beta transus during upsetting
Finish	20 to 50	1700	Should yield proper proportion of equiaxed alpha
			Rib-and-Web Shape
Draw and flatten	10 to 80	1750-1900	Higher temperature allows for longer time at forging temperature; embrittled structure will be removed by subsequent operations
Block	40 to 70	1650	Should provide adequate deformation in areas that exceeded beta transus during previous operation
Finish	30 to 50	1700	Should yield proper proportion of equiaxed alpha

(a) Temperatures given represent metal temperature; higher furnace settings are sometimes used to provide faster heating rates.

For parts receiving small total reductions (less than 50 per cent), it is desirable to begin with a billet that has a structure that contains at least 10 to 15 per cent of equiaxed alpha. This is especially true of parts that are upset forged from billets having length/diameter ratios over 1/2. Such parts contain large "dead-metal" zones which receive little or no work.

Beta Alloys. The general influence of forging temperature on the room-temperature mechanical properties of the Ti-13V-11Cr-3Al beta alloy is illustrated in Figure 14-10. Characteristic of other single-phase alloy systems, the beta-titanium

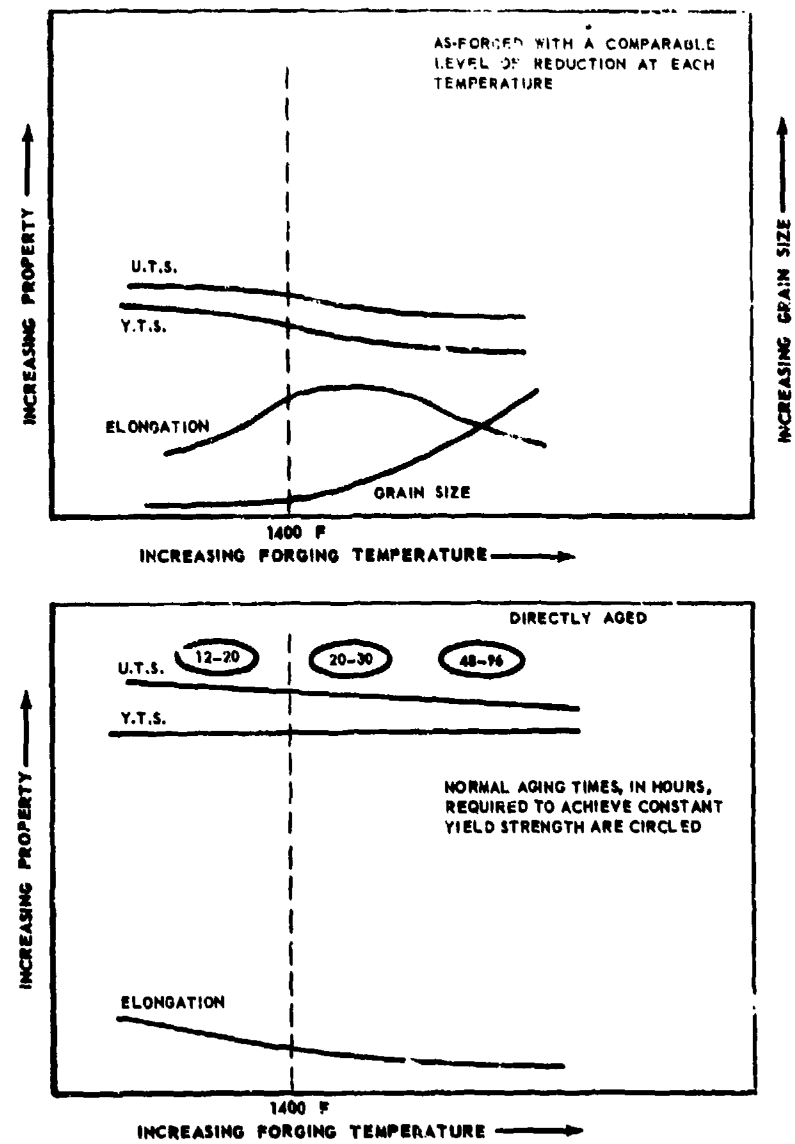


FIGURE 14-10. GENERALIZED INFLUENCE OF FORGING TEMPERATURE ON ROOM-TEMPERATURE MECHANICAL PROPERTIES OF Ti-13V-11Cr-3Al BETA ALLOY IN THE AS-FORGED AND AGED CONDITIONS

Typical effect of forging temperature on grain size shown for as-forged condition

alloy exhibits decreasing ductility with increasing grain size, and increasing strength with increasing amounts of cold work. Although the alloy recrystallizes at about 1400 F, cold work is retained after forging at temperatures as high as 1600 F. When cold work is retained, precipitation of the alpha phase occurs more rapidly and evenly during subsequent aging. This shortens the aging time needed to achieve a given strength level with good ductility. After forging at temperatures above about 1600 F, longer aging times are required, and the alpha precipitate concentrates at the grain boundaries, causing low ductility. These points are demonstrated by the aging-response curves in Figure 14-11, which show more rapid aging with increasing forging reductions and with decreasing forging temperature. The aging response is slowest after annealing.

The beta alloy has a potential yield strength capability exceeding 200,000 psi. However, at these higher strengths, the alloy is quite brittle and exhibits elongation values consistently below 5 per cent. For this reason, most investigations have concentrated on achieving a yield-strength level of about 180,000 psi, where higher elongation values may be expected. Even at this strength level, careful control of forging temperatures and reductions is important for controlling aging response and, hence, mechanical properties.

The influence of various forging sequences on the mechanical properties of the Ti-13V-11Cr-3Al alloy are as follows.[9]

(1) Fastest aging response and highest ductilities are obtained when the alloy is warm forged at or slightly below the recrystallization temperature (about 1400 F).

(2) The aging reaction is hastened by increasing amounts of reduction in this temperature range.

(3) The ductility of annealed forgings increases with decreasing grain size.

(4) The ductility of aged forgings is more dependent on prior working conditions than on grain size.

(5) The finest, most uniform grain size is obtained by forging in multiple steps with successively lower forging temperatures. Since grain growth is rapid above 1500 F, final forging reductions are usually performed in the vicinity of this temperature to obtain the optimum combination of grain refinement and aging response.

A summary of various forging techniques for the beta alloy is shown in Figure 14-12 which gives forging reductions, temperatures, aging cycles, and the resulting values for yield strength and elongation. The data for forging sequences I and J give properties for die forgings finished at 1500 F and aged for 12 and 15 hours. The aging cycle giving the highest elongations was 12 hours at 900 F.

Comparisons were made by Brody and Crosby of the properties of all-beta disk forgings produced by both hammer and press-forging techniques.[10] While the properties generally were comparable, the hammer-forged parts exhibited the lowest strength in the central regions because of "dead-metal" zones there. The press-forged parts received more uniform deformation, hence more uniform properties were obtained.

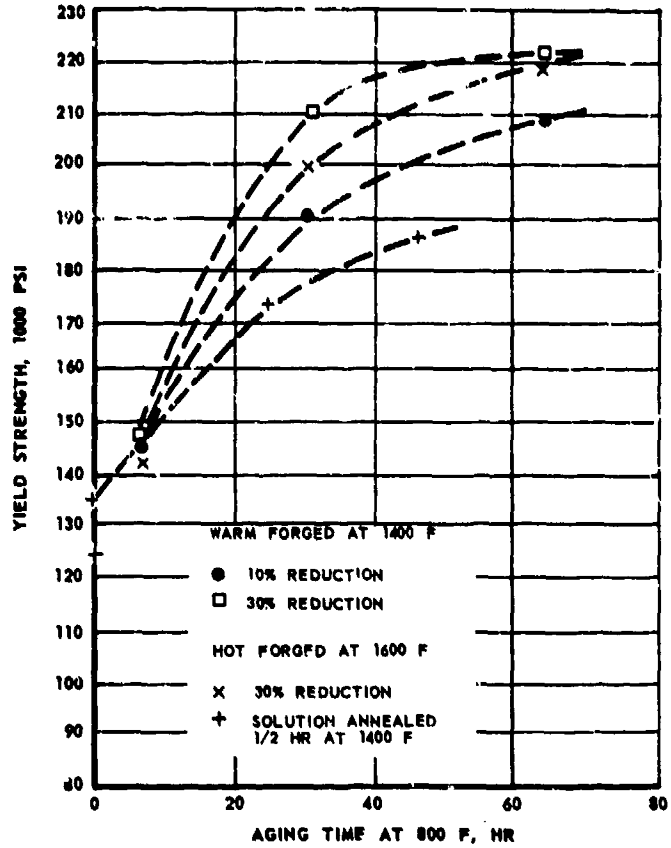


FIGURE 14-11. AGING RESPONSE OF Ti 13V-11Cr-3Al ALLOY FORGINGS AT 800 F AFTER VARIOUS AMOUNTS OF COLD WORK(4)

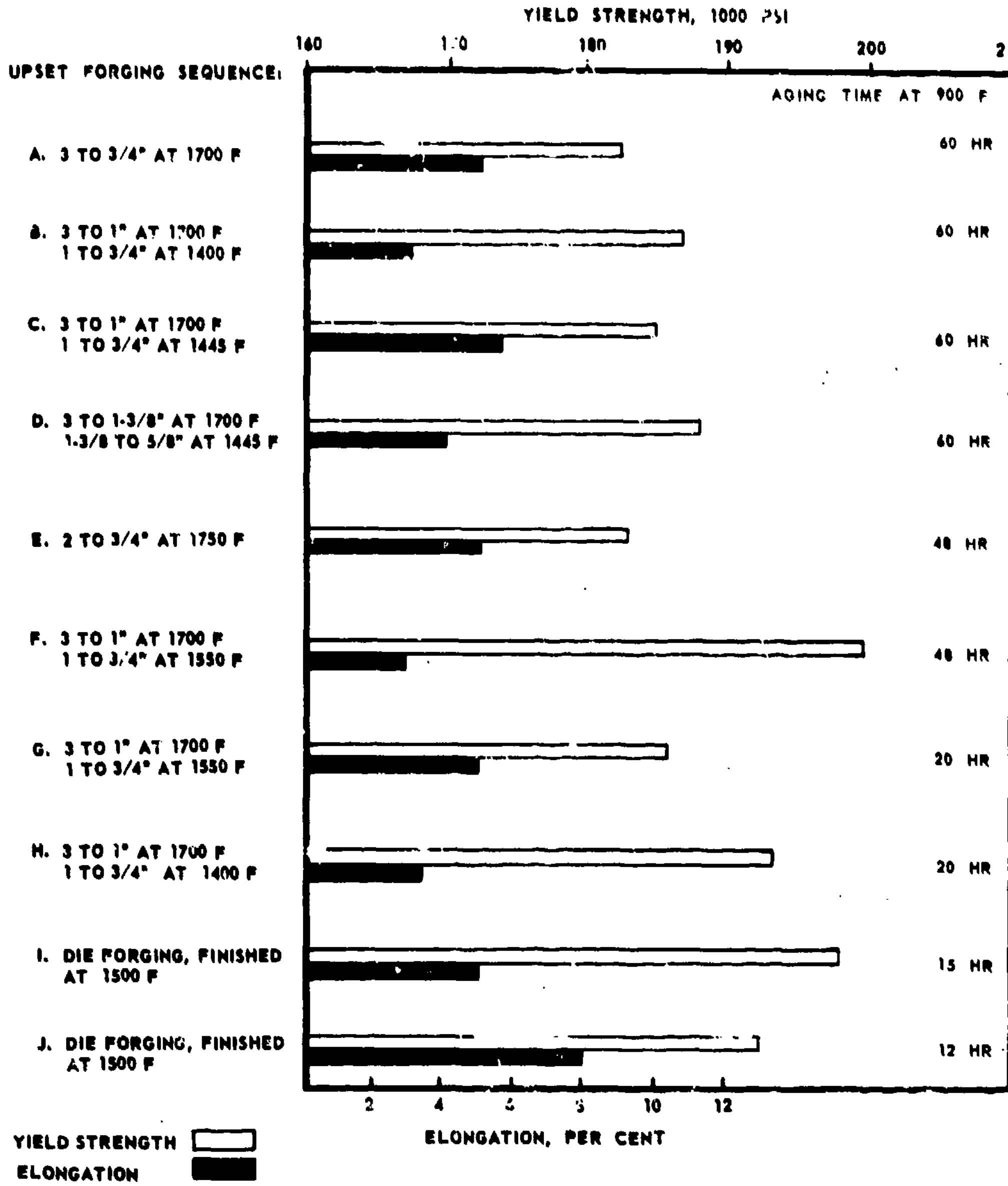


FIGURE 14-12. INFLUENCE OF VARIOUS UPSET FORGING SEQUENCES ON THE YIELD STRENGTH AND ELONGATION OF THE Ti-13V-11Cr-3Al BETA ALLOY[9]

Alpha Alloys. The general influence of forging temperatures on the mechanical properties of alpha alloys is illustrated in Figure 14-13. The behavior is similar in many respects to that for the alpha-beta alloys. The strength level is not influenced greatly by forging temperature, but the ductility decrease sharply when forging temperatures exceed the beta transus. The best combinations of forgeability and mechanical properties are at forging temperatures within the alpha-beta field. Since this represents a narrower forging temperature range than for the alpha-beta alloys, the alpha alloys are usually reheated more often during forging.

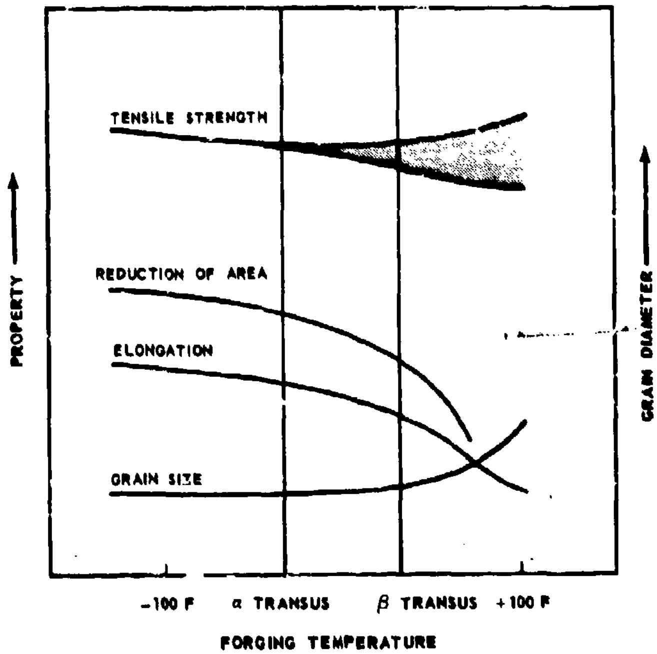


FIGURE 14-13. GENERALIZED INFLUENCE OF FORGING TEMPERATURE ON GRAIN SIZE AND ROOM-TEMPERATURE MECHANICAL PROPERTIES OF FORGED ALPHA ALLOYS

Certain of the alpha alloys (e.g., Ti-8Al-1Mo-1V) exhibit improved creep resistance when forged above the beta transus. However, the problem of increasing grain size introduces ductility variations that are difficult to control. The typical influence of forging temperature on the room-temperature mechanical properties of the Ti-8Al-1Mo-1V alloy is shown by the tabulation below for samples upset forged with 75 per cent reduction:

Forging Temperature[a], F	0.2% Offset Yield Strength, ksi	Ultimate Strength, ksi	Elongation, per cent	Reduction of Area, per cent
1850	128-139	140-155	14-15	21-31
1950	114-123	131-142	10-15	22-24

Data courtesy of Titanium Metals Corporation of America.

(a) After forging, samples annealed 1 hour at 1650 F and 8 hours at 1100 F.

COMMENTARY ON FORGING PRACTICES

There are a number of factors that make titanium alloys more difficult to forge than steels. The metallurgical behavior of the alloys not only imposes certain controls and limitations on the forging operation, but influences all of the steps in manufacturing forged parts. Particular care is necessary throughout the processing cycle to avoid contamination by oxygen, nitrogen, carbon, and/or hydrogen, which can severely impair the properties and over-all quality of a forged part.

Composition Sensitivity. The forging behavior and mechanical properties of individual titanium alloys are not very sensitive to variations in metallic alloying elements. However, both of these factors are influenced significantly by variations in interstitial elements (e.g., oxygen, nitrogen, and carbon). An increase from 0.1 to 0.2 per cent oxygen, for example, will raise the beta transus of the Ti-6Al-4V alloy as much as 75 F and increase the strength level by as much as 15 per cent. The typical influence of oxygen content on mechanical properties is illustrated for two forging alloys by the following data:[4,11]

Nominal Oxygen Content, %	Typical Tensile Strength, psi	
	Alpha-Beta Ti-6Al-4V (Annealed)	Beta Ti-13V-11Cr-3Al (Aged)
0.10	135,000	190,000
0.15	147,000	205,000
0.20	155,000	214,000

Increasing carbon and nitrogen contents cause similar but less-noticeable changes in strength.

Heating. At temperatures above about 1000 F, oxygen and nitrogen react with titanium to form an adherent surface scale and a hard, alpha-rich subsurface layer. The subsurface layer is brittle even at forging temperatures and can cause rupturing during forging. The alpha alloys are particularly sensitive to this and forgings of these alloys sometimes require frequent in-process grinding operations. When the beta alloy is forged above about 1800 F, it exhibits the same characteristics.

When hydrogen is absorbed at forging temperatures, it diffuses inward, raising the hydrogen content of the entire forging. In extreme cases, this can lead to hydrogen embrittlement. Thus, when titanium alloys are heated for forging in conventional oil- or gas-fired furnaces, oxidizing atmospheres are preferred to minimize hydrogen pickup. Inert-gas atmospheres are recommended only for parts containing extremely thin sections that require multiple heating operations.

Removal of Contaminants. The hard, alpha-rich subsurface layer produced by heating titanium in air is difficult to machine. Depending on heating times the layer ordinarily ranges in thickness from 0.005 to 0.025 inch. Pickling forgings in hydrofluoric acid (2-4 per cent aqueous solution) will remove the layer at a rate of about 0.001 inch per minute. To minimize hydrogen contamination from the acid bath, the solution is usually modified with about 30 per cent nitric acid. This retards the metal removal rate by about 50 per cent. Hydrogen may be removed only by vacuum annealing. Titanium alloys are easily analyzed for hydrogen content, and appropriate quality-control measures can be readily established for maintaining hydrogen contents below harmful levels.

Lubrication. The adherent scale formed on titanium billets during heating is abrasive and causes dies to wear at rates considerably faster than occurs with comparable steel forgings. The use of glass-type billet coatings for titanium alloys has reduced die wear to about the same level as for stainless steels. The glass-type coatings provide a twofold advantage of reducing oxidation and providing lubrication. Other common die lubricants are colloidal graphite and the soot from partially burned kerosene. Those lubricants are often applied by spraying, especially for parts forged in intricate die cavities. Swabs are used for simple shapes like disks and blades.

Equipment Cleanliness. While forging titanium, care should be taken to prevent contact with steel scale. A "thermite" type reaction can occur which can ruin a forging die. Apparently, the titanium reduces iron oxide in an exothermic reaction set off by pressure and high temperature. This reaction is known in forge shops as "firing". Furnace hearths and the forging equipment may be sources for steel scale.

Straightening. Because of the low elastic modulus and relatively high strength of titanium alloys, forgings are difficult to cold straighten either by coining or reverse bending. Such operations are usually done at temperatures between 700 and 1000 F. At times, it is helpful to maintain a straightening load on a forging for a few seconds to take advantage of "creep". This technique is especially useful for removing large warpages. Creep-resistant alloys (e.g., Ti-5Al-2.5Sn, Ti-8Al-1Mo-1V) require straightening temperatures upwards of 1000 F, especially if straightening requirements are severe. Forgings having severely contaminated surfaces are apt to crack during straightening. Hence, contamination should be removed beforehand.

Forging Technique. The successful forging of titanium alloys, particularly the alpha-beta types, depends a great deal on the forge-hammer operator and the choice of equipment capacity. No matter how carefully the dies are designed and the furnaces are controlled, the operator can cause beta embrittlement by forging a part too rapidly.

When forging disk shapes, for example, it is often desirable to allow one "free-swing" of the rams before each hammer blow. This allows time for the workpiece temperature to equalize. It is also desirable to avoid "over-hammering" or using larger hammers than necessary for the part. Some companies will intentionally "under-hammer" for initial breakdown forging operations so that temperature increases will not be significant. For example, a disk part normally requiring a 15,000-lb hammer might be blocked in a 12,000-lb hammer with more blows and then finished in a 15,000-lb hammer.

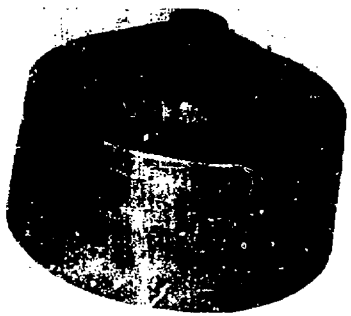
COMMENTARY ON FORGING DESIGN

Design principles for alpha and alpha-beta titanium alloys fall between those for alloy steels and hot-cold work alloys (e.g., 16-25-6 steel, unalloyed molybdenum).* The designs should provide an adequate amount of deformation during forging, but the reductions need not be confined to narrow limits. Because the beta alloy requires precise control of reductions, the designs are limited to generously contoured shapes. The size of titanium forging is limited only by the equipment available throughout the industry. Disk shapes as large as 70 inches in diameter and intricate rib-and-web parts over 4 feet long have been forged successfully. Some examples of typical parts forged in production quantities are shown in Figure 14-14.

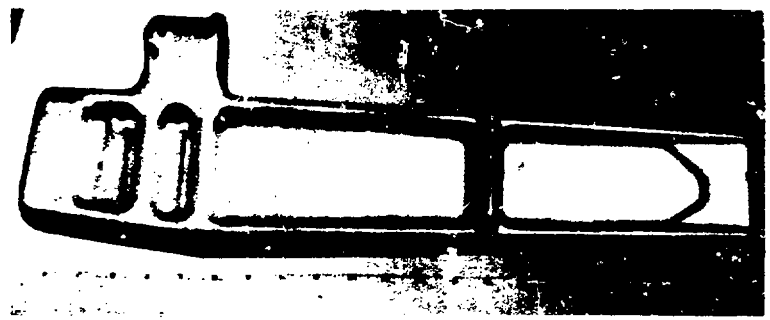
Section Size Limitations. Forgings containing massive sections greater than about 3 inches thick generally exhibit decreasing strength properties with increasing thickness. This is due primarily to slower cooling rates. The influence on alpha-beta alloys is greater. It is possible, however, to reduce large sections by machining or slotting and to restore properties by subsequent heat treatment. Forgings containing both massive and thin sections exhibit a greater tendency for hydrogen contamination because the entire forging requires longer heating times. In general, designs containing sections larger than 3 inches thick should not contain sections smaller than about 1 inch thick. For the heat-treatable alpha-beta alloys, it is important to avoid severe changes in section size to maintain a uniform heat treatment response. These alloys exhibit optimum heat treatment response in section thicknesses smaller than 1 inch.(4)

Tolerances. Producibility of close-tolerance, precision titanium-alloy forgings has been proven.(12) However, owing to such factors as excessive die wear, the need for expensive tooling, problems of microstructure control and contamination, the cost of close-tolerance forging is usually prohibitive. Metallurgical quality is also sometimes compromised if several dies are needed to accomplish the forging. Successful precision forging, therefore, is confined to small forgings, such as blades and fittings, that do not have complex flow patterns. An example of a close-tolerance blade forging is shown in Figure 14-15. Close-tolerance specifications should be avoided on rib and web designs, since the metal flow in these designs usually precludes the use of several dies.

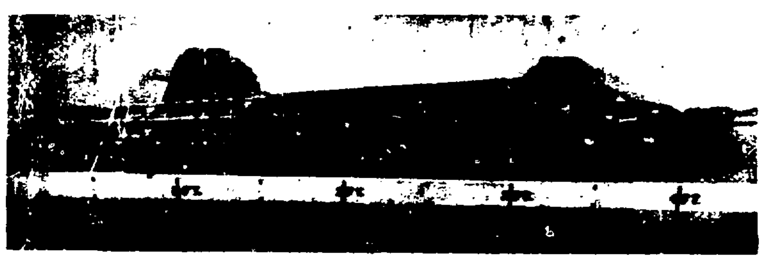
*Forging designs for titanium alloys are described in considerable detail in Chapter 4.



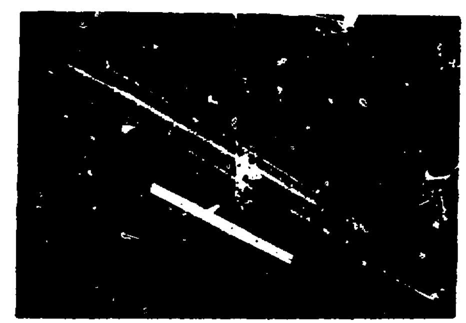
a. Ti-6Al-4V Head, 15-in. Diameter, 60 lb



b. Ti-6Al-4V Engine-Mount Aft-Beam, 15 lb



c. Ti-6Al-4V Beam, 46 lb



d. Ti-7Al-4Mo Spar, 34 lb

FIGURE 14-14. TYPICAL PARTS FORGED TO CONVENTIONAL DESIGNS FROM HEAT-TREATABLE TITANIUM ALLOYS

Courtesy Steel Improvement and Forge Company and Wyman-Gordon Company

FIGURE 14-15. CLOSE-TOLERANCE FORGING OF Ti-6Al-4V ALLOY BLADES

Forging steps: (1) extrude and upset, (2) block, (3) finish and trim, and (4) coin

REFERENCES

(1) Henning, H. J., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", Technical Documentary Report No. ML-TDR-64-95, Contract No. AF 33(600)-42963, Battelle Memorial Institute (March 1964).

(2) Griest, A. J., Sabroff, A. M., and Frost, P. D., "Effect of Strain Rate and Temperature on the Compressive Flow Stresses of Three Titanium Alloys", Trans. ASM, 51, 935-944 (1959).

(3) Southern Research Institute, "Short-Time Elevated Temperature Mechanical Properties of Metals Under Various Conditions of Heat Treatment, Heating and Exposure Time and Loading", Summary Report, Army Ballistic Missile Agency, Contract DA-01-009-ORD-494 (July 17, 1959).

(4) Henning, H. J., and Frost, P. D., "Titanium-Alloy Forgings", DMIC Report No. 141, Defense Metals Information Center, Battelle Memorial Institute (Dec. 19, 1960).

(5) Colton, R. M., "The Effects of Fabrication Procedure on the Properties of Ti-6Al-6V-2Sn - 0.5Fe-0.25Cu", Watertown Arsenal Laboratories Monograph Series (Sept. 14-15, 1959).

(6) Hamer, J. E., "Investigation to Develop Optimum Properties in Forged Ti-7Al-4Mo", WADD TR 60-489, Crucible Steel Company, Contract AF 33(616)-6122 (October, 1960).

(7) Croan, L. S., and Rizzitano, F. J., "The Influence of Forging Temperature on Mechanical Properties of Al-V Titanium Alloys", Trans. AIME, 51, 999-1014 (1959).

(8) Griest, A. J., Young, A. P., and Frost, P. D., "Relation Between Beta Grain Size and Ductility of High Strength Alpha-Beta Titanium Alloy", Trans. AIME, 215, 844-848 (October, 1959).

(9) Erbin, E. F., and Broadwell, R. G., "Heat Treatment of Forgings and Heavy Bar Sections of the Ti-13V-11Cr-3Al All-Beta Alloy", Titanium Metals Corporation of America. Paper presented at Sixth Annual Titanium Metallurgical Conference, N. Y. U. (September, 1960).

(10) Brody, R. P., and Crosby, F. A., "Research and Development of Titanium Rocket Motor Cases", Pratt & Whitney Aircraft, Division of United Aircraft Corporation, Progress Reports on Contract DA-19-020-ORD-5230.

(11) Hicks, J. S., "Establishment of Production Methods for Aircraft Precision Forgings in Titanium Alloys", North American Aviation, Los Angeles, Final Engineering Report to AIME, Contract AF 33(600)-33703 (November, 1959).

CHAPTER 15

IRON-BASE SUPERALLOYS

Crystalline Structure:	Face-centered cubic
Phase Changes:	None
Number of Phases:	One at forging temperature
Solidus Temperature Range:	2250 to 2500 F
Reactions When Heated in Air:	Oxidize slowly

ALLOYS AND FORMS AVAILABLE

Iron-base superalloys generally comprise a class of austenitic stainless steels that contain about 25 to 35 per cent nickel, 15 per cent chromium, and smaller amounts of molybdenum, tungsten, and other elements. There are exceptions to this, however; one alloy (19-9DL) is a modification of 18-8 stainless. Nominal compositions of several typical iron-base superalloys are given in Table 15-1. Alloys of this type have been used extensively as forgings for turbine and compressor components in jet engines. Because they contain reactive elements (e.g., titanium, aluminum), these alloys are commonly melted under vacuum in either arc or induction furnaces to avoid oxidation losses. Forging stock is commonly prepared by consumable-electrode melting either in vacuum or under a slag cover (Kellogg process). This is done to assure a high degree of cleanliness and uniform forgeability.

TABLE 15-1. NOMINAL COMPOSITIONS OF SEVERAL IRON-BASE SUPERALLOYS

Alloy	Composition, weight per cent										
	C	Mn	Si	Cr	Ni	Mo	Ti	Al	W	Others	Fe
A-286	0.08	1.35	0.50	15.0	26.0	1.25	2.0	0.25	--	0.3V	Bal
V-57	0.08[(a)]	0.25[(a)]	0.55	15.0	25.5	1.25	3.0	0.25	--	0.3V 0.008B	Bal
M-308	0.08	--	--	14.0	32.5	4.0	2.2	0.30	6.5	0.004B 0.25Zr	Bal
W-545	0.05	1.5	0.40	13.5	26.0	1.5	2.8	0.20	--	0.08B	Bal
Discaloy	0.06	1.0	0.70	14.0	26.0	3.0	1.7	0.25	--	--	Bal
16-25-6	0.06	1.35	0.70	16.0	25.0	6.0	--	--	-	0.15N	Bal
19-9DL	0.30	1.10	0.6	19.0	9.0	1.25	0.3	--	1.25	0.4Cb+Ta	Bal

(a) Indicates maximum.

Iron-base superalloys derive their strength by one or more of the following mechanisms: (1) solid-solution hardening, (2) precipitation hardening, and (3) work hardening. The basic strength of iron is increased by additions of soluble elements (e.g., chromium, nickel, molybdenum, etc.), each of the elements contributing to the

ultimate strength level of the usual solid solution. Since the elements are completely dissolved and form a homogeneous single-phase alloy, they do not affect forgeability to any great extent. However, these elements raise the recrystallization temperatures, thus increasing the minimum hot-working temperatures over that of iron.

The precipitation hardening of austenitic iron-base alloys occurs by the precipitation of intermetallic compounds and phases formed from alloy additions like aluminum and titanium — elements that are not completely soluble in iron. These compounds and phases exhibit increasing solubility with increasing temperatures and dissolve completely at temperatures approaching 1800 F, thus permitting the classical strengthening by solution heat treating and aging. In the A-236 alloy, the chief hardening constituents are the intermetallic compounds Ni_3Al and $Ni_3(Al,Ti)$, which precipitate at temperatures in the vicinity of 1300 F. Some carbide precipitation may also occur at the aging temperature and may contribute slightly to the strengthening of the steels.

Until about 1955, it was believed that boron, if present in the alloys, contributed little if at all to the strength. Since that time, it has become evident that boron plays an important role in the precipitation of the Ni_3Al compound.[1] In the absence of boron, this compound precipitates in a coarse, lamellar fashion rather than as a finely divided, uniform precipitation. The alloy is significantly weakened when this occurs. Lamellar precipitation is accelerated by prior cold work, particularly when solution-heat-treated forgings are cold coined or straightened prior to aging. If the alloys contain small amounts (0.001 per cent or more) of boron, this problem is virtually eliminated.

Since the 19-9DL and 16-25-6 alloys contain essentially no precipitation hardeners, their high recrystallization temperature permits the alloys to be strengthened by hot-cold working at temperatures between about 1200 F and 1700 F. Hot-cold working improves the elevated-temperature strengths and stress-rupture properties of these alloys for service at temperatures up to about 1000 F. By comparison, the heat-treatable alloys such as A-286 retain equivalent strength levels and stress-rupture properties in the temperature range from about 1200 F to 1400 F.

Billets for forging are generally furnished as press-forged squares or as hot-rolled rounds depending on size requirements. Some forging companies use as-cast ingots for die forging.

FORGING BEHAVIOR

Cleanliness of the iron-base superalloys has probably the greatest influence on hot forgeability. The alloys containing reactive elements (e.g., titanium and aluminum) can develop nitride and carbonitride segregation that later show up as stringers in wrought bars, especially in air-melted heats. These stringers cause poor forgeability on billets upset forged under the best of conditions. The pancake forgings in Figure 15-1 illustrate the relative forgeability of air-melted and vacuum-melted A-286 materials. The appearance of stringers in the microstructure of air-melted material is illustrated by the typical structures of the forgings.

Temperatures and strain rates also have an important influence on forgeability of iron-base superalloys. Experiments at Battelle[5] have shown that the A-286 alloy is quite forgeable between 1400 F and 2200 F at pressing speeds of 1.5 inches per second. When forged at about 250 inches per second, however, the alloy shatters at 1400 F and exhibits evidence of hot shortness at 2200 F. Between 1600 F and 2000 F, forgeabilities

750X

Air Melted

Vacuum Melted

750X

FIGURE 15-1. EFFECT OF MELTING PRACTICE ON THE MICROSTRUCTURE AND FORGEABILITY OF A-286 ALLOY(2)

Poor forgeability of air-melted alloy is associated with nitride stringers.

at these two rates were comparable. Thus, strain rate or deformation rate tends to narrow the hot-working temperature range for the A-286 and similar alloys.

Below about 1600 F, most of the iron-base superalloys exhibit precipitation reactions that reduce forgeability. This is reflected by a marked drop in elevated-temperature ductility. Figure 15-2 compares the tensile elongations at elevated temperatures of Discaloy and A-286 with two austenitic stainless steels.[3] Such graphs are useful for predicting forging temperatures. However, it is believed that hot impact-tensile tests* give a truer index of the relative ductility at temperatures in the hot-working range and at strain rates comparable to forging rates. Figure 15-3 shows how the ductility of the V-57 alloy, as measured by hot impact-tensile tests, drops sharply in the precipitation temperature range.[4] These data indicate that the optimum forging temperature for V-57 is between 1700 F and 2000 F.

Because of their high aging temperatures, the iron-base superalloys retain high strengths at comparatively high temperatures. Figure 15-4 compares the typical elevated-temperature tensile strengths of two of these superalloys with that of Type 304 stainless steel. Such curves, however, do not adequately reflect the forging-pressure requirements for these alloys. Upset forging studies at Battelle showed that forging pressures increase as much as 50 per cent with increasing reductions, even at temperatures of 2000 to 2200 F. As shown in Figure 15-5, the actual forging pressures may be from two to four times greater than the tensile strengths at corresponding temperatures.

Strain rates or deformation rates have important influences on the forging pressures of these steels. Figure 15-6 shows that the forging pressure for A-286 alloy increases significantly with a 10-fold increase in strain rate that is typical of the rates common to presses. Similarly the specific energy required for upsetting increases with increasing rate. It is shown in Figure 15-7 that increasing energy is required for increasing strain rates (encompassing both presses and hammers) at each temperature, and that the percentage increase becomes more significant at the higher temperatures.

For all practical purposes, the iron-base superalloys exhibit essentially the same level of forging pressure. At forging temperatures they require slightly greater forging pressures than the 18-8 stainless grades. However, because the forging temperature range is narrower, forging companies usually allow for 10-20 per cent larger equipment capacity.

COMMENTARY ON FORGING PRACTICES

During the early 1950's, the die forging of iron-base superalloys was complicated by many intermediate steps needed for removing ruptures caused by segregation in the ingots. Since that time, segregation has been virtually eliminated through advanced melting techniques. The alloys may now be forged in a greater variety of configurations with heavier reductions, approaching the forgeability of austenitic stainless steels. The rib-and-web forging shown in Figure 15-8, for example, illustrates the complex parts made recently from vacuum-melted A-286. To forge such parts from the earlier air-melted alloys would have been impractical because of the many painstaking intermediate operations normally required.[6] Figure 15-9 shows a landing arrestor hook also forged from the A-286 alloy. Thus, it can be seen that many advances can be made toward improving the forging quality of alloys that may, at one time, be considered extremely difficult to forge.

*The hot impact-tensile test for forgeability is described in Chapter 6.

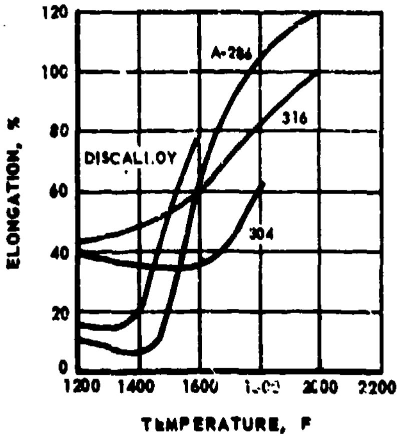
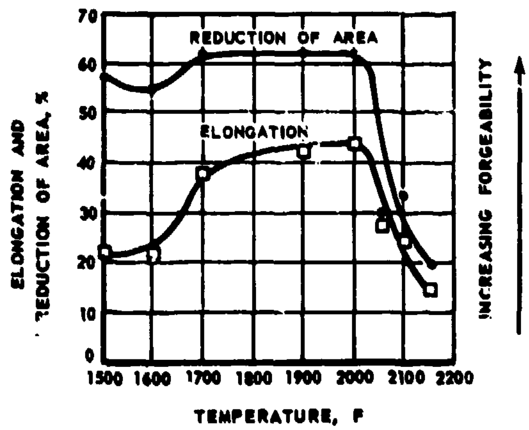


FIGURE 15-2.

COMPARATIVE ELEVATED-TEMPERATURE DUCTILITY OF IRON-BASE SUPERALLOYS AND AUSTENITIC STAINLESS STEELS[3]

FIGURE 15-3.

DUCTILITY OF V-57 ALLOY IN HOT IMPACT-TENSILE TESTS INDICATES AN OPTIMUM FORGING-TEMPERATURE RANGE OF 1700-2000 F[4]



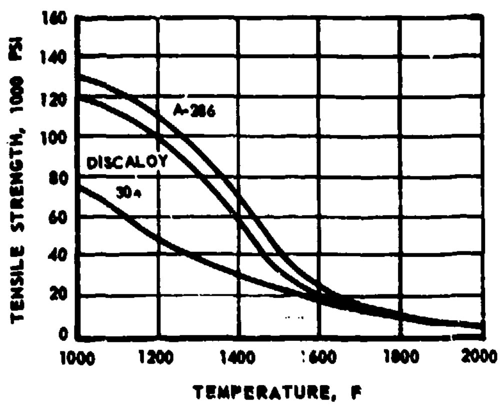


FIGURE 15-4.

COMPARATIVE TENSILE STRENGTH OF IRON-BASE SUPERALLOYS AND TYPE 304 STAINLESS STEEL AT ELEVATED-TEMPERATURE[3]

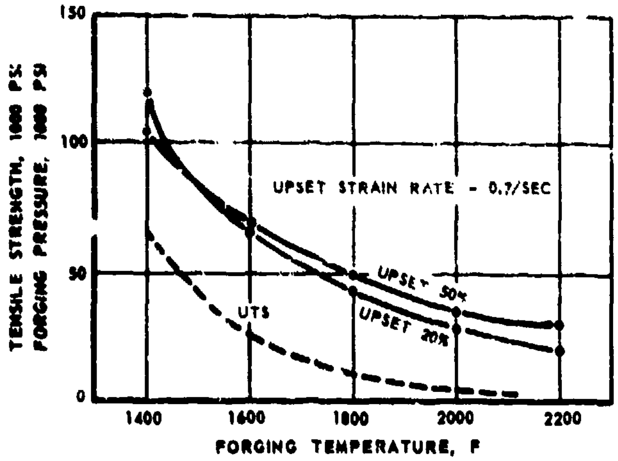


FIGURE 15-5. COMPARISON OF UPSET FORGING PRESSURE AND TENSILE STRENGTH OF A-286 ALLOY AT ELEVATED TEMPERATURES(3,5)

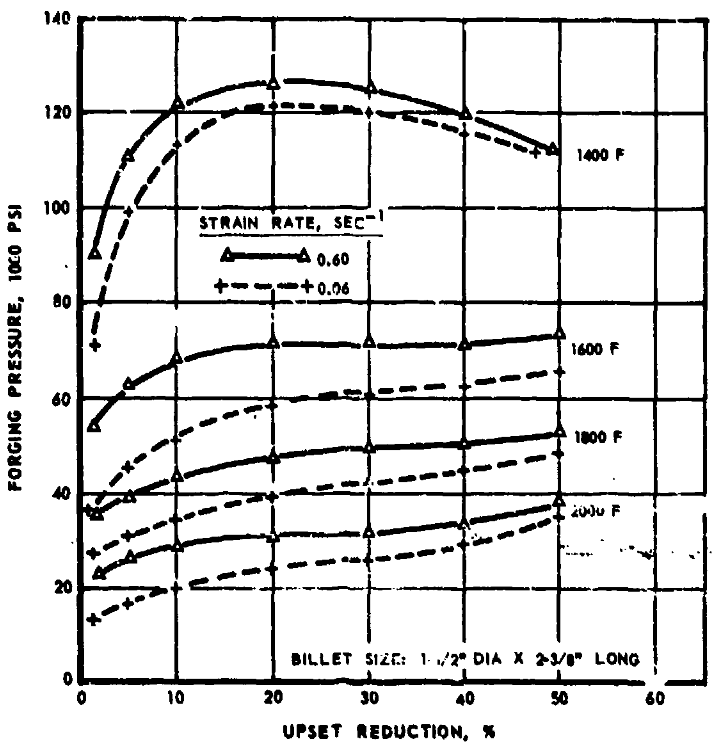


FIGURE 15-6. FORGING-PRESSURE CURVES FOR A-286 ALLOY UPSET FORGED AT VARIOUS TEMPERATURES AND TWO NOMINAL STRAIN RATES(5)

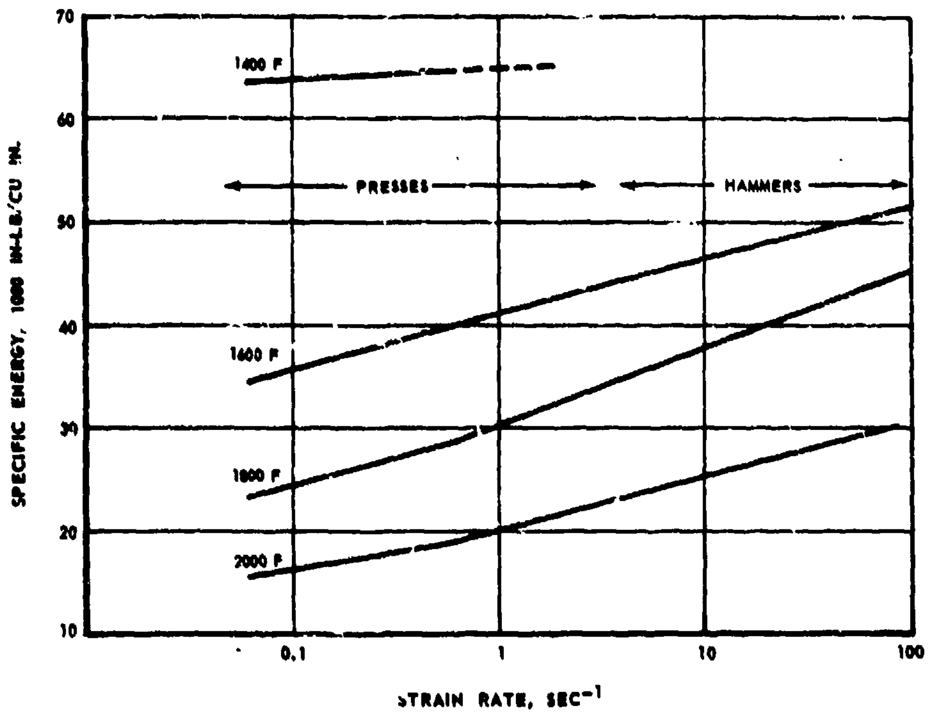


FIGURE 15-7. INFLUENCE OF STRAIN RATE AND TEMPERATURE ON THE SPECIFIC ENERGY REQUIRED FOR UPSET FORGING A-286 ALLOY 50 PER CENT[5]

FIGURE 15-8. COMPLEX A-286 ALLOY RIB-AND-WEB FORGING

Courtesy Canton Drop Forging and Mfg. Co.

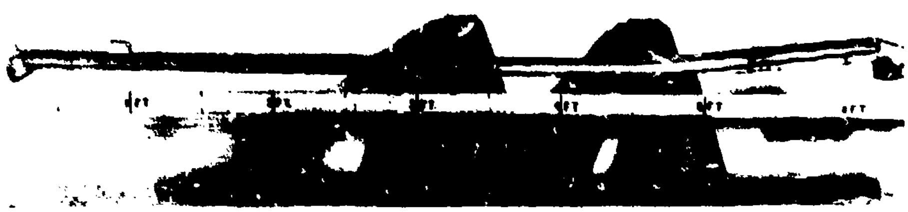


FIGURE 15-9. FORGED A-286 ALLOY LANDING ARRESTOR HOOK

Courtesy Steel Improvement and Forge Co.

Selection of Forging Temperatures. When selecting the forging temperatures for these alloys, it is important to consider such factors as the amount and rate of reduction, and the starting billet condition (as-cast or wrought). Since the alloys do not transform, grain size is controlled entirely by the amount and temperature of reduction preceding the solution-annealing treatments. The A-286 alloy, for example, can be forged at temperatures as high as 2200 F. In practice, however, forging temperatures between 1950 F and 2050 F are preferred to avoid possible incipient melting and to prevent excessive grain growth. Table 15-2 lists grain sizes for the A-286 alloy after forging at various temperatures and at reductions from 10 to 60 per cent, both before and after heat treatment. These data show how the as-forged grain size increases with increasing forging temperature and decreases with increasing reductions. The grain size of forgings of these alloys is also affected by any subsequent annealing. In this case, the temperature and degree of forging and the temperature and time of annealing govern the grain size obtained. Actually, the intermediate forging temperature of 1900 F gives the most uniform annealed grain size although not necessarily the smallest.

TABLE 15-2. INFLUENCE OF FORGING TEMPERATURE AND REDUCTION ON THE GRAIN SIZE OF A-286 ALLOY IN THE AS-FORGED AND ANNEALED CONDITIONS

ASTM grain size of as-received bar = 7.5 to 8.

Forging Temperature, F	Condition of Sample	ASTM Grain Size at Given Per Cent Reduction[a, b]			
		10	20	40	60
1800	As forged	7.5, 8	8, 6	8, 6	9.5, 8
	1950 F, 1/2 hr	4, 2	4, 3	4, 4.5	5.5, 4.5
1900	As forged	6, 5.5	6, 5.5	7, 5	5, 6
	1950 F, 1/2 hr	3.5, 4.5	4.5, 5	4, 5	4.5, 3.5
2000	As forged	3, 2	3, 2.5	3, 4.5	4, 4.5
	1950 F, 1/2 hr	3, 2	3	3.5, 4.5	5, 4.5

(a) First numerical rating indicates predominating grain size; second number indicates that there were a number of grains of this size.

(b) Ratings represent average values for three observations made by two observers.

As-cast ingots of iron-base superalloys generally contain coarse columnar grains that cause poor forgeability. For turbine and compressor disks, such materials cannot ordinarily be forged directly with large reduction. Rather, opposed conical preform dies are often used to impact small but uniform strains throughout the ingot (see Chapter 6). Temperatures for this operation are generally lower than normal to insure that the strained metal will recrystallize to a finer grain size during subsequent heating operations. It is sometimes helpful to anneal the preform at the maximum forging temperatures (2050 F for A-286), before continuing the forging operation. This hastens the recrystallization process and permits uniform forgeability.

For wrought materials forging temperatures are normally adjusted according to the amount of reduction the forging receives. Recommended forging temperatures for several precipitation-hardenable iron-base superalloys are given in Table 15-3.

The hot-cold work grades (i.e., 19-9DL, 16-25-6) are normally finish forged with 25 to 50 per cent reduction at temperatures between 1200 F and 1700 F, and air cooled. This is done ordinarily to achieve 30 to 40 per cent higher yield strengths along with satisfactory creep-rupture strength at the maximum operating temperatures than if

fully annealed. Typical room-temperature strengths and hardnesses for hot-cold forged 16-25-6 alloy are given below:

Condition	Tensile Strength, psi Ultimate	0.2% Yield	Hardness, Brinell
Annealed at 2100 F	120,000	82,000	207
Hot-cold forged 25% at 1700 F	138,000	111,000	279
Hot-cold forged 23% at 1200 F	162,000	143,000	326

These data show that much higher strengths can be achieved when this alloy is forged at lower temperatures with comparable levels of reduction. Ordinarily, the 16-25-6 and 19-9DL alloys are hot-cold finish forged in the vicinity of 1500 F with reductions of about 25 per cent.

TABLE 15-3. FORGING TEMPERATURES RECOMMENDED FOR SEVERAL WROUGHT SUPERALLOYS

Alloy	Maximum Forging Temperature, F	Recommended Forging Temperature for Indicated Reductions, F Light (Up to 15%)	Moderate (15-50%)	Large[a] (Over 50%)	Variable[b]
A-286	2200	1950	2000	2050	1950
V-57	2150	1950	2000	2050	2000
Discaloy	2200	1900	2000	2050	1900
M-308	2150	--	2000	2050	--

(a) If forgings undergo large reductions at rapid deformation rates (e.g., in mechanical presses or high-energy-rate machines), 50 to 100 F lower temperatures are recommended.

(b) This refers to forgings containing widely differing reductions. End upsets, for example, undergo sizeable reductions on the head end while the shaft end remains essentially undeformed.

Heat Treatment. The precipitation-hardenable iron-base superalloys must be heat treated after forging to exhibit maximum strength. The heat treatment usually consists of an anneal at about 1900 to 1950 F followed by an aging treatment at about 1200 to 1300 F. Such heat treatment is normally performed in an air atmosphere furnace. Air cooling after heat treating is recommended, although oil or water quenching may be employed after annealing heavy sections. Heat treatment in a so-called reducing atmosphere will not prevent scaling and may cause contamination from carbon or nitrogen. Furthermore, the scale formed by nonoxidizing environments is particularly difficult to remove by pickling.

REFERENCES

(1) Brown, H., "Metallurgical Characteristics of A-286 Alloys", DMIC Memorandum No. 59, Defense Metals Information Center, Battelle Memorial Institute (July 26, 1960).

(2) Dyrkacz, W. W., "Quality Improvements in Stainless Steels and Superalloys by Vacuum Melting", Paper presented at ASM Western Metals Congress, Los Angeles, California (March, 1959).

(3) Metals Handbook, 8th Edition, Vol 1: Properties and Selection of Materials, American Society for Metals, Copyright 1961.

(4) Data courtesy of Carpenter Steel Company, Reading, Pennsylvania.

(5) Henning, H. J., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", Technical Documentary Report No. ML-TDR-64-95, Contract No. AF 33(600)-42963, Battelle Memorial Institute (March, 1964).

(6) Henning, H. J., and Boulger, F. W., "Superalloy Forgings", DMIC Memorandum No. 86, Defense Metals Information Center, Battelle Memorial Institute (February 10, 1961).

CHAPTER 16

COBALT-BASE SUPERALLOYS

Crystalline Structure:	Face-centered cubic
Phase Changes:	None
Number of Phases:	One at forging temperature
Solidus Temperature Range:	2350 to 2650 F
Reactions When Heated in Air:	Oxidizes slowly

ALLOYS AND FORMS AVAILABLE

The physical metallurgy of cobalt-base alloys is similar in many respects to that of the iron-base superalloys.[1] Solid-solution hardening can be achieved by such alloy additions as chromium, tungsten, molybdenum, and columbium. Also, the alloys can be precipitation hardened by the addition of titanium. Nominal compositions of several important wrought cobalt-base superalloys are given in Table 16-1. Included are data useful for predicting forging behavior of each alloy.

TABLE 16-1. NOMINAL COMPOSITIONS AND FORGING CHARACTERISTICS OF SEVERAL COBALT-BASE SUPERALLOYS

Alloy Designation	V-36	S-816	L-605(HA-25)	J-1570	J-1650
Nominal Composition, %	0.30C	0.38C	0.10C	0.20C	0.20C
	25Cr	20Cr	20Cr	20Cr	19Cr
	20Ni	20Ni	10Ni	28Ni	26Ni
	4.0Mo	4.0Mo	15W	7.0W	12W
	2.0W	4.0W	1.0Fe	4.0Ti	2.0Cb
	2.0Cb	4.0Cb	1.5Mn		4.0Ti
	3.0Fe	3.0Fe			0.02B
	1.0Mn	1.2Mn			
Forging Temperature, F					
Range	1600-2250	1600-2250	1600-2300	1800-2200	1950-2200
Recommended	2200	2150	2250	2150	[illegible]
Relative Forgeability					
Presses	--	Good	Fair	--	--
Hammers	Good	Good	Good	Fair	Fair

Although cobalt-base alloys are strengthened at elevated temperatures by solid-solution hardening and carbide precipitation, their strengths may be increased even further by either hot-cold working or precipitation of the intermetallic compound Ni_3Ti.[1,2] From the standpoint of forging applications, the ability to strengthen the alloys by precipitation hardening is an advantage since hot-cold working is difficult to control in forged shapes. Thus, the V-36, S-816, and L-605 alloys are ordinarily hot forged and subsequently annealed. The heat-treatable J-1570 and J-1650 alloys are solution annealed and aged in a manner similar to the iron-base superalloys.

The solid-solution and carbide-precipitation-hardened alloys are generally melted with practices similar to those for austenitic stainless steel. The precipitation-hardenable alloys containing titanium, however, are ordinarily vacuum melted. Billets are usually supplied in wrought forms for subsequent forging. Cobalt-base alloys are not widely used in the form of forgings except in the case of blades and buckets for turbojet applications. While cobalt-base alloys, as a group, possess elevated-temperature strengths comparable to nickel-base alloys, the latter group has been more widely used for large forgings.

FORGING BEHAVIOR

Clark[3] has pointed out that the requirements related to service conditions and hot processing are diametrically opposed as far as wrought superalloys are concerned. In the one case, it is desired to have as high a degree of strength as possible at the operating temperature, whereas at the other, a suitable degree of plasticity is required at the processing temperature. In the cobalt-base superalloys this problem is complicated by the fact that high alloy contents, which impart good properties, often decrease their permissible hot-working range, so that processing and operating temperatures are brought closer together.

Many of the cobalt-base alloys cannot be forged successfully. Compared with the iron-base superalloys, the cobalt-base superalloys ordinarily contain more carbon and, therefore, greater quantities of hard metallic carbides which lead to forgeability problems. Of the alloys which can be wrought (Table 16-1), the forging behavior can be classified into two groups according to the hardening mechanism:

> Solid-Solution and Carbide-Precipitation-Hardened Alloys. The V-36, S-816, and L-605 alloys exhibit fairly good forgeability over a rather wide range of temperatures from about 1600 F to 2300 F. In this respect the alloys are comparable to iron-base superalloys such as 19-9DL and 16-25-6, but the forging pressures are on the order of three to four times greater for the cobalt-base alloys.

> Precipitation-Hardened Alloys. The J-1570 and J-1650 alloys also exhibit fairly good forgeability, but their forging temperature range is narrower — between about 1800 F and 2200 F. Outside this range the forgeability of the alloys is limited either by the onset of precipitation or by the formation of low-melting constituents. In this respect, these alloys compare with such iron-base superalloys as A-286 and M-308. The forging pressures required are similar to the other cobalt-base alloys.

Figure 16-1 compares the elevated-temperature strength and ductility of several wrought cobalt-base superalloys with the A-286 iron-base superalloy. Here it can be seen that the cobalt-base alloys are considerably stronger and far less ductile than the A-286 alloy at forging temperatures. These data also show how the ductility (and the forgeability) of the J-1650 alloy markedly improves above 1800 F where precipitated intermetallic compounds begin to dissolve.

COMMENTARY ON FORGING PRACTICES

Of the cobalt-base superalloys, those most widely used in the form of forgings are S-816 and L-605. Even when forged at their upper forging temperatures, these alloys exhibit work-hardening behavior; thus, forging-pressure requirements increase with increasing reductions. Accordingly, the alloys generally require frequent reheating during forging. This promotes recrystallization and lowers the forging pressure for succeeding steps. Surface cracks occurring during forging are generally removed mechanically between these forging steps.

Forging conditions (temperature and reduction) have an important effect on the grain size of cobalt-base superalloys. Since many characteristics, particularly low ductility, notch brittleness, and low fatigue strength are associated with coarse grains, the proper control of working and final heat treatment is very important.

Selection of Forging Temperatures. Cobalt-base superalloys are subject to grain growth when heated to temperatures above about 2150 F. The alloys heat rather slowly and require long soaking times for temperature uniformity. Forging temperatures (and reductions), therefore, depend on the type of forging operation and the part configuration.

The alloys are usually forged with small reductions in initial breakdown operations. The reductions are selected to impart sufficient strain to the metal so that recrystallization (and usually grain refinement) will occur during subsequent reheating. Since the partially forged section size is smaller, shorter times are necessary to reach temperature uniformity, and reheating temperatures are sometimes 50 F to 150 F higher than for the initial forging operation. If the part receives only small reductions in the subsequent forging steps, however, it is important to continue forging at the lower temperatures. These small reductions must exceed about 5 per cent, however, to avoid abnormal grain growth during subsequent annealing.(6) Forging temperature recommended for two types of parts reflect these considerations:

Typical Operations	Recommended Forging Temperatures, F S-816	L-605(HA-25)	J-1570
	Disk Forging		
Upset No. 1	2150	2150	2100
Upset No. 2	2250	2250	2200
Block	2200	2200	2150
Finish	2200	2200	2150
	Disk with Integral Shaft		
Draw one end	2150	2150	2100
Roll shaft	2250	2250	2150
Block	2150	2150	2100
Finish	2150	2150	2100
(Anneal)	(2150)	(2200)	(2150)

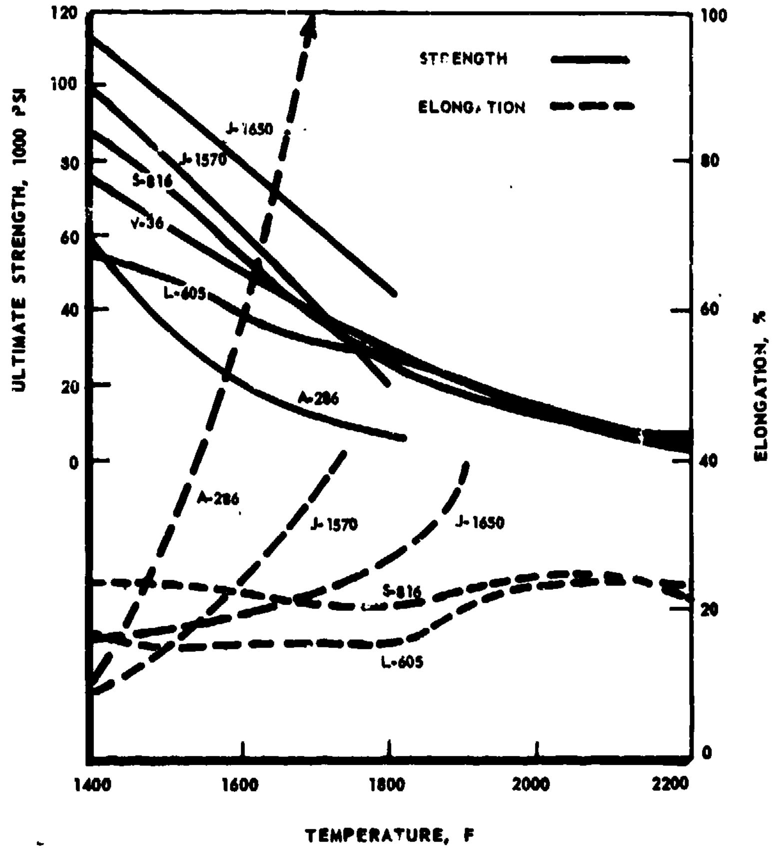


FIGURE 16-1. ELEVATED-TEMPERATURE STRENGTH AND DUCTILITY OF WROUGHT COBALT-BASE SUPERALLOYS(4,5)

Data for A-286 iron-base superalloy shown for comparison

In forging the disk with an integral shaft, ordinarily the deformation given to the shaft is complete after the rolling operation. Thus, the shaft remains undeformed while the disk portion is being worked. In this final operation, the lower forging temperature is important for preventing grain growth in the shaft. Under these circumstances, it is good practice to maintain subsequent forging temperatures below the annealing temperatures given in parentheses.

Control of Reduction. The final grain size obtained in wrought cobalt-base superalloys is controlled by the amount of reduction. Forgings containing regions that receive "critical" reductions (between 1 and 5 per cent) are subject to abnormal grain growth upon subsequent annealing. The preferred practice consists of forging with reductions of greater than about 20 per cent to avoid such regions. It is worth noting that forgings that are restruck in finish dies are particularly sensitive to abnormal grain growth.

Design. Large forgings from cobalt-base alloys are generally designed with generous contours and with a minimum of detail. Blades, on the other hand, may be forged to thin sections as long as the blades contain no abrupt changes in section size.

REFERENCES

(1) Cobalt Monograph, Centre d'Information du Cobalt, Brussels, Belgium (1960).

(2) Wagner, H. J., and Hall, A. M., "The Physical Metallurgy of Cobalt-Base Superalloys", DMIC Report 171, Defense Metals Information Center, Battelle Memorial Institute (July 6, 1962).

(3) Clark, C. L., High-Temperature Alloys, Pitman, New York (1953).

(4) "Alloy Digest Nos. Co-3, Co-5 and Co-11", published by Engineering Alloys Digest, Inc., Upper Montclair, N. J.

(5) Metals Handbook, 8th Ed., Vol 1: Properties and Selection of Metals, Amer. Soc. Metals, Cleveland, Ohio (1961).

(6) Decker, R. F., Rush, A. I., Dano, A. G., and Freeman, J. W., "Abnormal Grain Growth in M-252 and S-816 Alloys", NACA Tech. Note 4084 (November, 1957).

CHAPTER 17

NICKEL-BASE SUPERALLOYS

Crystalline Structure:	Face-centered cubic
Phase Changes:	None
Number of Phases:	One phase (gamma) at forging temperatures. On cooling, certain phases precipitate from matrix
Solidus Temperature Range:	2250 to 2450 F
Reactions When Heated in Air:	Oxidize slowly

ALLOYS AND FORMS AVAILABLE

Nickel-base superalloys comprise a family of compositions that has grown remarkably in the past decade. Initially, the alloys consisted of a few, rather simple nickel-chromium alloys hardened by small additions of titanium and aluminum. These alloys were useful for service temperatures up to about 1400 F. With the development of production vacuum-melting techniques, it has become possible to produce workable alloys containing relatively large amounts of titanium, aluminum, zirconium, columbium, and other reactive elements. The levels of gaseous elements (e.g., nitrogen and oxygen) are reduced by vacuum melting; this eliminates most of the nitrides and oxides that otherwise contribute to poor workability. Hence, the second generation of nickel-base superalloys consists of numerous compositions containing larger amounts of hardening elements. Such alloys as Waspaloy, René 41, Astroloy, and Udimet 700 have extended the service-temperature range upwards to 1800 F, offering useful strengths closer to the melting range than most other alloy systems.

Table 17-1 lists nominal compositions for several commonly forged nickel-base superalloys. The alloys having the highest useful service temperatures are generally those containing highest levels of aluminum, titanium, and other reactive elements. They also are the most difficult to forge. These alloys are strengthened by three basic mechanisms:

(1) Solid-solution hardening

(2) Precipitation hardening

(3) Complex hardening derived from small boron and zirconium additions.

It is important to review these mechanisms because they play important roles in the forging behavior of superalloys.

TABLE 17-1. NOMINAL COMPOSITION OF SEVERAL WROUGHT NICKEL-BASE SUPERALLOYS

Alloy Designation	Nominal Composition, weight per cent												
	C	Cr	Ni	Mo	Al	Ti	Co	Fe	B	Zr	Mn	Si	Others
Hastelloy Alloy W	0.10	5.0	Bal	25.0	--	--	1.5	5.0	--	--	0.5	0.5	0.25 V
Nimonic Alloy 80 A	0.10	20.5	Bal	--	0.20	0.35	--	0.5	--	--	0.5	0.5	--
Hastelloy Alloy X	0.10	22.0	Bal	9.0	--	--	1.5	18.5	--	--	0.5	0.5	0.6 W
Inconel Alloy 718	0.04	19.0	52.5	3.[illegible]	0.6	0.8	--	18.0	--	--	0.2	0.2	5.2 Cb
Incoloy Alloy 901	0.05	13.5	42.7	6.2	0.25	2.5	1.0	34.0	Trace	--	0.45	0.4	--
Inconel Alloy X-750	0.04	15.0	Bal	--	0.6	2.4	0.4	6.5	--	--	0.5	0.2	0.85 Cb
D-979	0.04	15.0	45.0	4.0	1.0	3.0	--	27.0	0.01	--	0.4	0.4	4.0 W
Hastelloy Alloy R-235	0.10	16.0	Bal	5.5	2.0	2.5	1.9	10.0	Trace	--	0.25	0.5	--
M-252	0.11	19.0	Bal	9.5	1.0	2.5	10.0	2.5	0.005	--	0.2	0.3	--
Waspaloy	0.06	19.5	Bal	4.2	1.2	3.0	13.5	1.0	0.08	0.09	0.5	0.4	--
René 41	0.09	19.0	Bal	9.6	1.5	[illegible].2	11.0	--	0.005	--	0.01	0.02	--
Inconel Alloy 700	0.12	15.0	Bal	3.8	3.0	2.2	28.5	0.7	--	--	0.1	0.3	--
Udimet 500	0.09	19.0	Bal	4.0	2.8	3.0	17.0	2.0	0.008	--	--	--	--
Udimet 700	0.09	15.0	Bal	5.2	4.2	3.5	18.5	0.5	0.008	--	--	--	--
Astroloy	0.06	15.5	Bal	5.3	4.5	3.6	15.5	0.2	0.030	--	0.05	0.3	--
Unitemp 1753	0.25	16.5	Bal	1.5	2.0	3.2	7.5	9.5	0.008	0.05	--	--	8.5 W

Solid-solution hardening of nickel is not different from that of other metals. The strength of nickel is increased by additions of soluble elements such as chromium, molybdenum, and tungsten. Each of these elements contributes to the strength of the final complex solid solution. Since the elements are completely dissolved and form a homogeneous, single-phase alloy, they do not affect the forgeability of the alloy to any great extent. However, the additions raise the recrystallization temperatures, thereby raising the minimum hot-working temperatures over that of nickel.

Hardening also results from the precipitation of compounds and phases formed from alloy additions of carbon, aluminum, titanium, and columbium, elements which are not completely soluble in the nickel. These compounds and phases exhibit increasing solubility with increasing temperatures. When temperatures are high enough, the compounds dissolve completely, allowing for the classical strengthening by solution heat treating and aging analogous to heat-treatable alloys like those of aluminum. The highly alloyed superalloys differ in precipitation behavior from that of aluminum in that several compounds and phases precipitate on aging. Also, the proportions and compositions vary, depending on the aging temperatures.

Figure 17-1, reproduced from studies on the René 41 alloy by Weisenberg and Morris[1], illustrates the relative abundance of each of the three important precipitating compounds, as influenced by different aging temperatures. This graph shows that the M_6C carbide will precipitate at temperatures up to about 2150 F. From the standpoint of elevated-temperature strength this is desirable. However, the carbides reduce forgeability especially if located at grain boundaries. Although the aging times indicated by Figure 17-1 are rather long for unstrained material, the precipitation reaction is hastened by plastic deformation somewhat analogous to the strain-induced precipitation occurring in the all-beta titanium alloy under certain conditions. Thus, when René 41 is deformed at temperatures just below 2150 F, the M_6C carbide causes reduced forgeability by precipitating during deformation. Precipitation of the other compounds, $Ni_3(Al, Ti)$ and $M_{23}C_6$, causes further reduced forgeability if the alloy is forged at correspondingly lower temperatures.

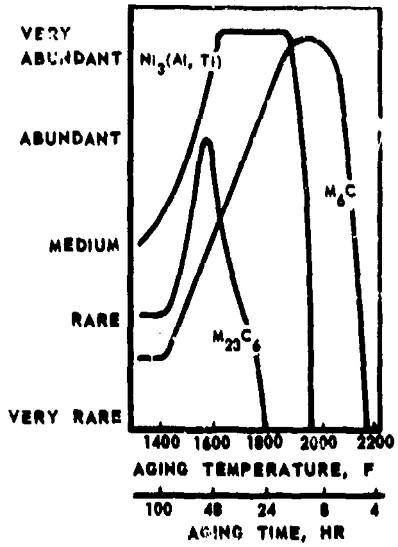


FIGURE 17-1.

ISOTHERMAL PHASE REACTIONS IN RENÉ 41[1]

Bar-stock specimens were initially treated at 2200 F and water quenched. After aging at the times and temperatures indicated, phases were extracted and examined by X-ray diffraction.

M represents the metallic radical (usually Cr, Mo, or possibly Ti)

[redacted]

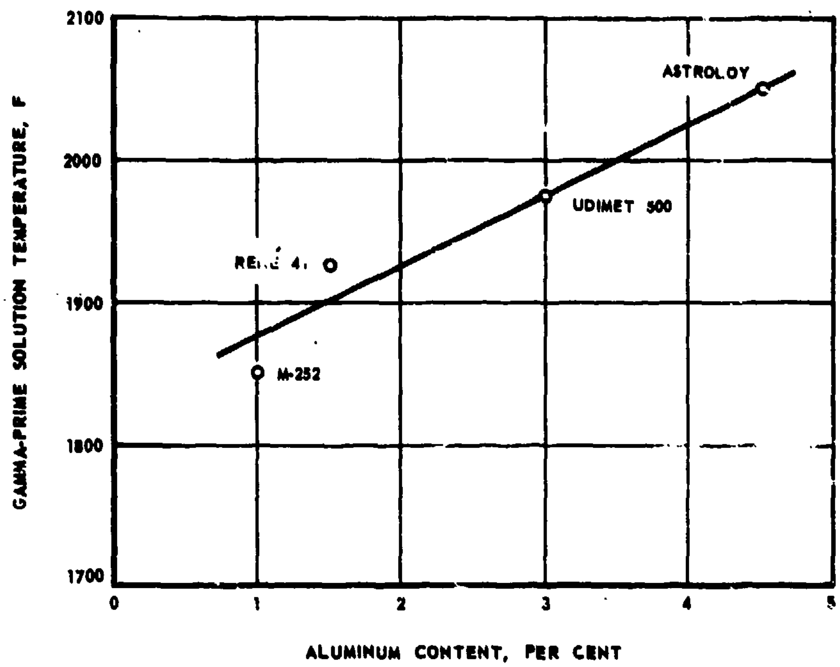


FIGURE 17-2. EFFECT OF ALUMINUM CONTENT ON THE GAMMA-PRIME $Ni_3(Al, Ti)$, SOLUTION TEMPERATURE[2]

Moon, et al.[a], have shown that the solution temperature for the gamma prime phase [$Ni_3(Al, Ti)$] increases with increasing aluminum content as shown in Figure 17-2. The solution temperature for this compound is about 1950 F for René 41; Udimet 500 and Astroloy exhibit even higher gamma-prime solution temperatures because of their higher aluminum contents. Alloys containing more than about 5 per cent aluminum are considered casting alloys and are not suitable for forging because they retain gamma-prime at temperatures very close to that of incipient melting.

Small amounts of boron and zirconium have been found to increase the creep resistance of nickel-base superalloys. It is believed that these elements fill the lattice vacancies normally present at or near the grain boundaries. Decker, et al.,[3] suggested that boron retards carbide formation at the grain boundaries by filling these vacancies. Zirconium is said to have a similar effect. If their plausible explanations are correct, small additions of boron or zirconium should improve forgeability by retarding the formation of embrittling carbides at grain boundaries; the reverse should be true when the amounts of these insoluble elements exceed those necessary to fill the vacancies.

Additions of cobalt have an opposite effect on forgeability from that of zirconium and boron. Cobalt tends to stabilize the gamma prime and carbides by raising their solvus temperatures in nickel-base alloys, thereby retaining the phases at even higher temperatures. Although this behavior is desirable from the strength standpoint, it impairs forgeability*.

For reasons related to elevated-temperature creep strength and weldability, alloys like René 41, Udimet 700, and Astroloy are often forged at temperatures below those where the M_6C carbide dissolves completely. It is believed that once the M_6C carbide goes into solution it does not reprecipitate in a fashion favorable for maximum creep strength. Rather, the compound changes to the lower temperature carbide form ($M_{23}C_6$) and precipitates at the grain boundaries. This reduces rupture ductility and stress-rupture strength to some extent and promotes sensitivity to thermal cracking. Thus, there are times when the selection of forging temperatures must be based on a compromise with factors contributing to final mechanical properties.

The nickel-base superalloys are available in a variety of cogged billet and bar sizes for forging. The alloys are ordinarily melted by one of the following techniques:

(1) Air melting followed by either vacuum-induction melting or vacuum consumable-electrode arc melting

(2) Vacuum-induction melting followed by vacuum consumable-electrode arc melting

(3) Consumable-electrode arc-melting under slag.

Compared with ordinary arc-melting techniques, these melting procedures have been responsible for remarkable improvements in forgeability and mechanical properties by reducing the levels of segregation. However, the majority of ingots produced on a production basis still contain certain amounts of segregation that influence forgeability. Since vacuum-induction melted ingots solidify progressively toward the center, and take longer to freeze, the alloys and impurities tend to concentrate at the center. The segregation in consumable-electrode arc-melted ingots is generally on a smaller scale

*Forgeability refers to the ability of a material to undergo plastic deformation without fracture, regardless of the forging pressure.

with greater concentration differences at the immediate ingot surfaces. For these reasons, wrought bars from consumable-electrode arc-melted ingots are generally more susceptible to edge cracking during forging than bars from vacuum-induction-melted ingots. Conversely, the latter are more susceptible to center cracking.

FORGING BEHAVIOR

As shown in Table 17-2, the temperature limits for forging nickel-base superalloys are determined largely by melting and precipitation reactions. The alloys appear in this table in the order of progressively narrower ranges between maximum and minimum forging temperatures, hence of progressively poorer forgeability. The estimated maximum forging-temperatures are considered low enough to avoid incipient melting during heating for forging.

TABLE 17-2. CRITICAL MELTING AND PRECIPITATION TEMPERATURES FOR SEVERAL NICKEL-BASE SUPERALLOYS(4,7)

	Approximate Temperature, F		Estimated Forging Temperature Range, F	
Alloy Designations	First Melting	Precipitation	Maximum	Minimum
Hastelloy Alloy X	2300	1400	2250	1500
Inconel Alloy 718	2300 (a)	1550	2200	1600
Waspaloy	2250 (a)	1800	2200	1700
Incoloy Alloy 901, AL 901, Altemp 1251, Udimet 200	2200	1800	2150	1750
Inconel Alloy X-750	2350 (a)	1750	2200	1900
M-252	2200 (a)	1850	2150	1750
Hastelloy Alloy R-235, Unitemp R235	2300	1900	2200	1850
René 41, Haynes Alloy No. R41, Altemp R41, Udimet R41	2250	1950	2200	1900
Udimet 500, Unitemp 500, Hastelloy Alloy 500, Allvac 500	2250	2000	2200	1950
Udimet 700	2250 (a)	2050	2200	1950
Astroloy	2250 (a)	2050	2200	2000

(a) Estimated from data provided by producers.

Another basis for selecting forging temperatures is the elevated-temperature ductility of the alloys. As shown in Figure 17-3, the tensile elongations for several nickel-base superalloys are lowest at temperatures coinciding with precipitation behavior and highest at temperatures corresponding with solutions of the precipitated compounds. These data also show that such Ni-Cr-Fe-Mo alloys as Inconel Alloy 718 and Hastelloy Alloy R-235 reach higher levels of ductility than do Ni-Cr-Co-Mo alloys such as René 41 or Udimet 500. The latter alloys also exhibit the poorer forgeability based on other criteria.

Approximate forgeability ratings for typical nickel-base superalloys are suggested in Table 17-3. The ratings are based on the span of the forging temperatures range, as limited by melting and precipitation, and on the ductility in hot-tensile tests at appropriate temperatures. On this scale of empirical ratings, the René 41 alloy would have about one-third the forgeability of Inconel Alloy 718. It should be noted that these ratings reflect only a relative ability to withstand deformation without failure. They do not indicate the energy or pressure requirements necessary for forging, nor can the ratings be related to low-alloy steels and other alloys that normally exhibit far better forgeability than these superalloys.

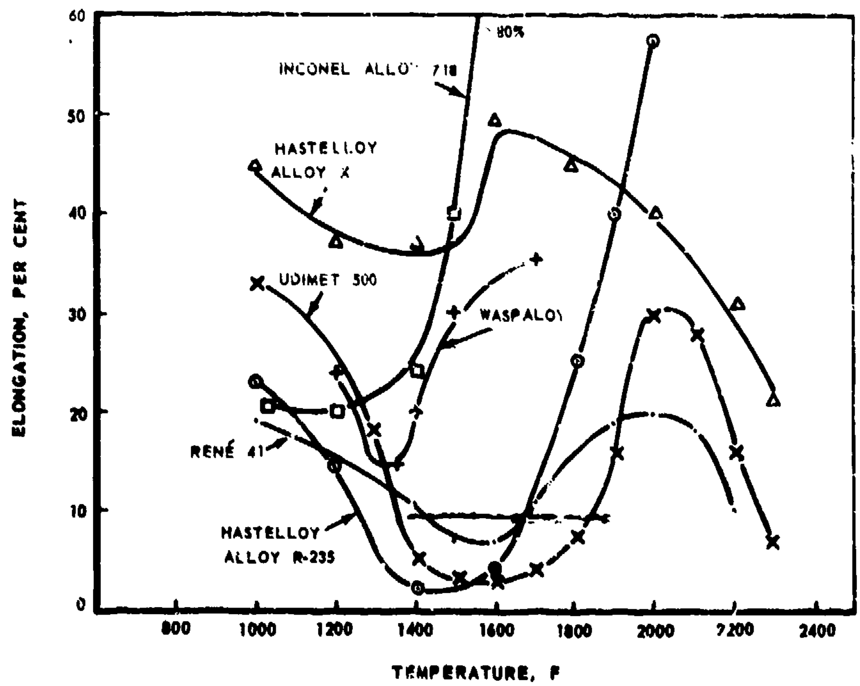


FIGURE 17-3. ELONGATION VALUES IN SHORT-TIME ELEVATED-TEMPERATURE TENSILE TESTS FOR SEVERAL NICKEL-BASE SUPERALLOYS[3,5,6,7]

TABLE 17-3. EMPIRICAL FORGEABILITY RATINGS FOR SEVERAL NICKEL-BASE SUPERALLOYS BASED ON LIMITING TEMPERATURES FOR FORGING AND TENSILE DUCTILITY

Alloy	Limiting Forging Temperatures, F (Based on Melting and Precipitation) Maximum	Minimum	Factor for Temperature Range(a)	Approximate Tensile Elongation at Forging Temperatures Temperature, F	Elongation, %	Factor for Ductility(b)	Forgeability Rating(c)
Hastelloy Alloy X	2250	1500	7.5	1900	40-45	4.5	12.0
Inconel Alloy 718	2200	1600	6.0	2000	>100	10	16.0
Inconel Alloy 901	2150	1750	4.0	1800	70-75	7.5	11.[illegible]
Waspaloy	2200	1750	4.5	2000	>40	5	9.5
Hastelloy Alloy R-235	2200	1850	3.5	2000	55-60	6	9.5
René 41	2200	1900	3.0	2000	15-20	2	5.0
Udimet 500	2200	1950	2.5	2000	25-30	3	5.5
Udimet 700	2200	1950	2.5	2000	25-30	3	5.[illegible]

(a) The factor for forging temperature is considered to be one-tenth the permissible range for forging as expressed in degrees Fahrenheit.

(b) The ductility factor is taken as one-tenth the elongation value in hot tensile tests at temperature near the middle of the forging range. The maximum value is 10 on this scale.

(c) Forgeability rating is the sum of factors for temperature range and ductility.

The forgeability of nickel-base superalloys usually decreases with increasing deformation rates, particularly for large reductions. The influence of rapid deformation rates and temperature on the forgeability of René 41 was studied at the Lockheed Aircraft Corporation.[8] The studies consisted of upsetting cylindrical samples at various temperatures ranging from 1600 F to 2200 F in a high-energy-rate forging machine (Model 1220 Dynapak). The samples shattered when upset forged below 2000 F and above 2200 F but not at intermediate temperatures. The influence of deformation rates was studied by increasing the Dynapak fire pressures. This change shortened the time during which deformation occurred and increased the amount of reduction obtained. Both of these factors contributed to increasing the metal temperature enough to cause cracking when the samples were upset much beyond 40-50 per cent in one stroke.

Figure 17-4 shows the combined influence of the rate and magnitude of deformation on the forgeability of René 41. The samples were upset at temperatures between 1950 F and 2100 F. The amounts and rates of reduction were controlled by varying the sample sizes and the ram speed. To obtain reductions greater than 40 per cent, it was necessary to use restriking steps of 40 per cent reduction and to reheat between operations.

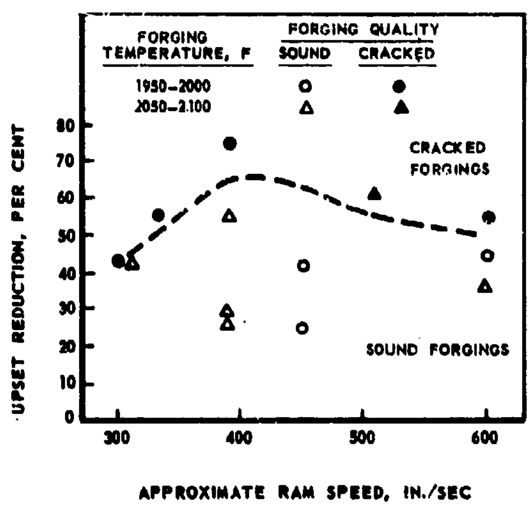


FIGURE 17-4. INFLUENCE OF REDUCTION AND RATE OF DEFORMATION ON THE FORGEABILITY OF RENÉ 41 FORGED AT TWO TEMPERATURES ON A HIGH-ENERGY-RATE MACHINE[8]

Another significant factor that reduces the forgeability of superalloys is chilling caused by forging between dies colder than the workpiece. Figure 17-5 shows how samples of René 41 appear after upset forging between colder dies. The severest ruptures occurred in the regions in contact with the dies not on the free surfaces. This

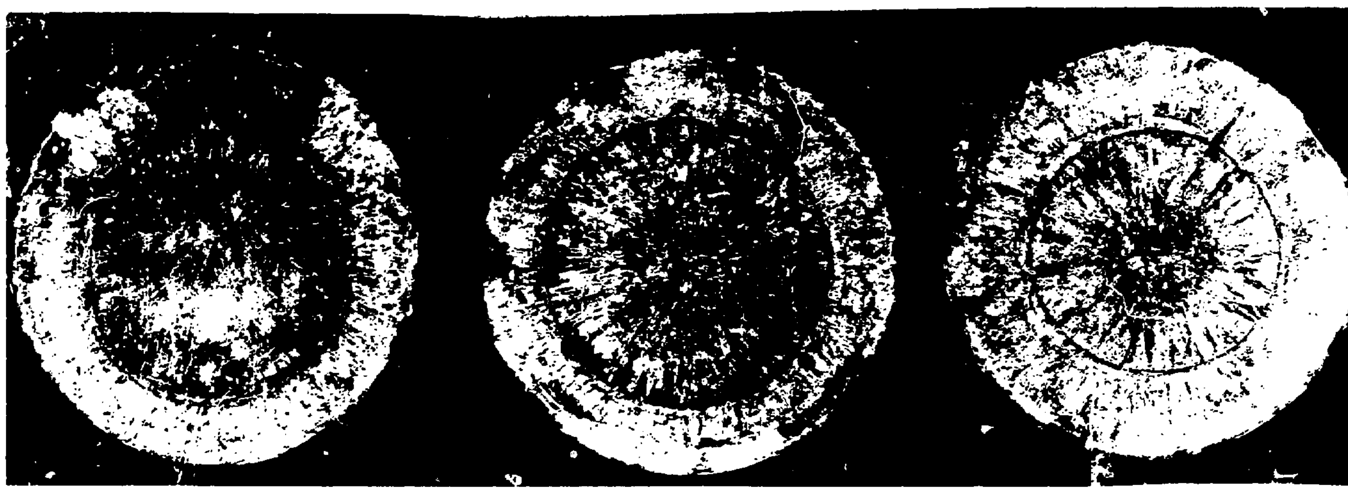


FIGURE 17-5. SAMPLES OF RENÉ 41 UPSET FORGED BETWEEN FLAT DIES SHOWING SEVERE CRACKING IN REGIONS IN CONTACT WITH DIES

One-half size
Courtesy of Wyman-Gordon Company

condition is most severe in forging alloys with the highest precipitation temperatures. Methods for preventing this type of cracking include the use of heated dies in combination with insulating coatings, like asbestos sheets, glass wool, or sawdust, to retard heat transfer from workpiece to dies. For obvious reasons, forgings containing thin, fast-cooling sections are extremely difficult to forge. One source reports benefits from placing sheet metal backed with asbestos between the dies and workpiece. Reportedly, the sheet metal helps to maintain a continuous layer of asbestos between the die and workpiece. Various forging companies have spent considerable time developing insulating materials. For this reason, the preferred techniques are not generally disclosed. It is believed, however, that the most successful methods include the use of glass coatings. Press-forging dies are probably heated in the vicinity of 1000 F, while hammer-die temperatures are probably closer to 400 F.

The forgeability behavior of the René 41 alloy is typical of the nickel-base superalloys in general. At slow rates, the alloys exhibit reasonable forgeability at temperatures as low as 1800 F. At the more rapid rates, however, the minimum temperature is about 1900 F. Thus, the widest temperature range for forging is obtained at low pressing speeds. Unfortunately, die-chilling offsets most of the advantages gained by forging at the slower rates.

Elevated-temperature strength data available in the literature provide information on the forging pressure requirements for various superalloys. Table 17-4 compares approximate yield strengths of several superalloys with those for AISI 4340 steel and the A-286 and 16-25-6 iron-base superalloys. These data show that, in general, the nickel-base superalloys will require greater forging pressures at working temperatures. This is particularly true for the alloys with high cobalt and Ti+Al contents.

TABLE 17-4. SHORT-TIME ELEVATED-TEMPERATURE YIELD STRENGTHS FOR SEVERAL SUPERALLOYS(9, 10)

Tested at Low Strain Rates

Alloy Designation	Approximate Yield Strength, ksi, at Indicated Temperatures, F			
	1600	1800	2000	2200
Nimonic Alloy 80A	30-35	4-6	--	--
Inconel Alloy 718	45-50	10-15	4-5	--
Incoloy Alloy 901	45-50	10-15	5-10	--
Hastelloy Alloy R-235	55-60	15-20	5-10	--
Hastelloy Alloy X	20-25	15-20	10-15	3-5
Waspaloy	65-70	20-25	5-10	--
René 41	55-60	20-25	10-15	--
M-252	60	35-40	--	--
Udimet 500	70-75	35-40	--	--
Astroloy	90-95	40-45	10-15	--
Udimet 700	90-95	40-45	15-20	--
4340	--	3-5	2-3	1-2
A-286	20-25	10-15	5-8	--
16-25-6	25-30	10-15	--	--

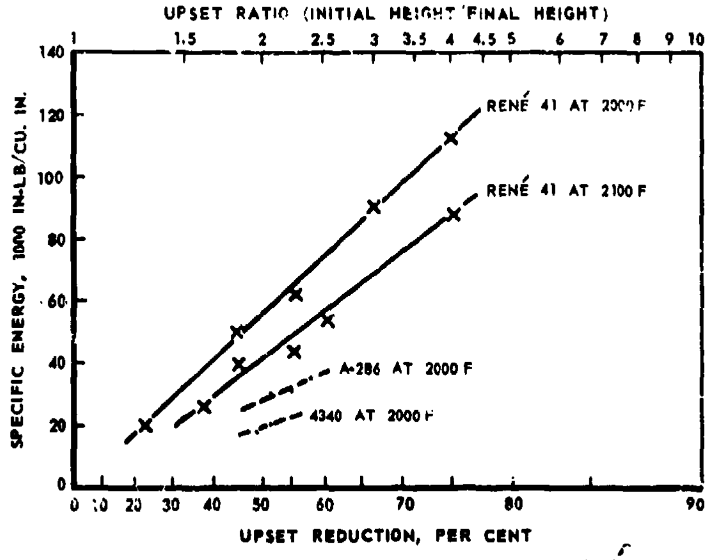


FIGURE 17-6. ENERGY REQUIREMENTS FOR UPSETTING RENÉ 41 ON A HIGH-ENERGY-RATE FORGING MACHINE[8]

Data for 4340 and A-286 included for comparison

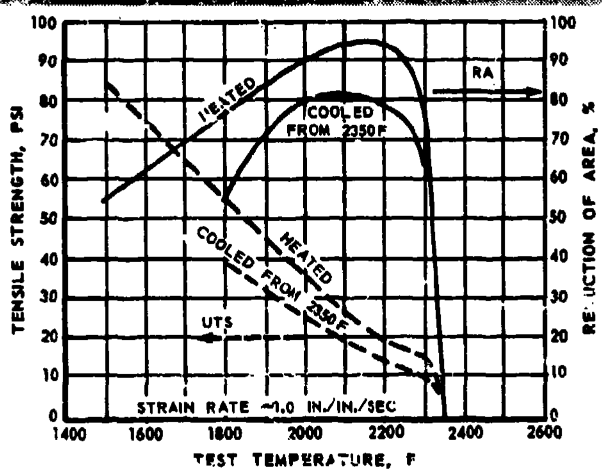


FIGURE 17-7. INFLUENCE OF TEMPERATURE ON TENSILE STRENGTH AND REDUCTION OF AREA OF INCONEL ALLOY X-750 ON DIRECT HEATING AND AFTER COOLING FROM 2350 F TO TESTING TEMPERATURE

After Nippes, et al.[11]

The relative magnitude of the actual forging pressures for the René 41, A-286, and 4340 alloys is indicated by the specific energy requirements for upsetting these materials to various reductions, shown in Figure 17-6. Specific energy data for René 41 alloy were developed from high-energy-rate forging studies conducted by Lockheed Aircraft[8]; similar data for A-286 and 4340 were obtained in studies at Battelle. Because of overheating and attendant fracture problems, the larger reductions on René 41 had to be obtained by incremental working with as many as four operations and with intermediate reheating steps. For a 50 per cent reduction, these data show that the forging pressure at 2000 F for René 41 will be at least twice that for A-286 and three times that for 4340. These data also bring out the importance of accurate temperature control with alloys such as René 41; approximately 20-25 per cent higher pressures are required for forging at 2000 F than at 2100 F.

Data from Nippes, et al.,[11] in Figure 17-7 show the influence of temperature on ductility and tensile strength of Inconel Alloy X-750 upon direct heating and upon cooling from 2350 F to the testing temperature. Both strength and ductility are reduced by superheating, since the higher preheating temperature causes both grain coarsening and more complete solution of carbides. For either heating practice, maximum ductility is obtained at temperatures in the vicinity of 2200 F where the tensile strength is 15,000-20,000 psi. These tests were made at comparatively rapid strain rates of about 1.0 in./in./sec. By comparison, at ordinary tensile-testing strain rates of 0.01 in./in./sec Inconel Alloy X-750 has a tensile strength in the vicinity of 7,000 psi at 2200 F. Thus, from the standpoint of forging pressures, advantage is gained by using the lowest strain rates practicable.

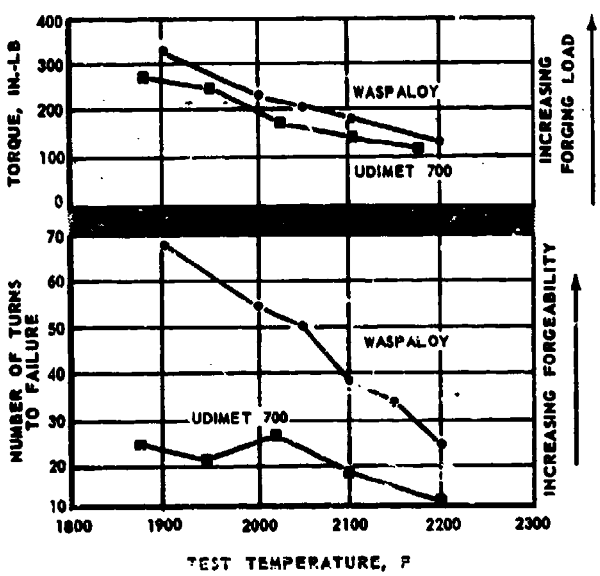


FIGURE 17-8. HOT-TWIST TEST DATA FOR WASPALOY AND UDIMET 700

[illegible] courtesy of Thompson-Ramo-Wooldridge

Although the various nickel-base superalloys exhibit a rather wide range of forgeability behaviors, the forging pressure requirements for most of the Ni-Cr-Co-Mo alloys are not too dissimilar. This is evidenced by the hot-twist data for Waspaloy and Udimet 700 in Figure 17-8. Waspalloy has markedly better forgeability than Udimet 700 over the forging temperature range, but the shear strength (as measured by the twisting torque) and, hence, the forging pressure requirements are quite similar. The temperatures for maximum twist ductility for these alloys agree quite closely with the forging temperatures in practice – about 1950 F for Waspaloy and about 2000 F for Udimet 700.

INFLUENCE OF FORGING VARIABLES ON MECHANICAL PROPERTIES

Factors influencing the mechanical properties of heat-treated nickel-base superalloys can be summarized as follows:

(1) The basic mechanical properties are obtained through appropriate heat treatments that are sometimes tailored to suit certain service requirements.

(2) Aside from differing heat-treatment procedures, variations of mechanical properties are caused by variations in cleanliness and chemical composition, precipitation and recrystallization behavior, grain size, grain flow, and amount of reduction.

Through deformation, forging has a significant influence on:

(1) Grain flow

(2) Recrystallization behavior

(3) Grain size

(4) Precipitation behavior.

Forging companies have expended considerable effort developing detailed forging procedures. Because of this, the procedures are considered proprietary and quantitative data relating forging techniques and mechanical properties are difficult to obtain. Consequently, the succeeding discussion contains limited amounts of mechanical-property data in support of the interpretations expressed.

Grain Flow. Superalloys exhibit anisotropic mechanical properties caused chiefly by carbide, nitride, and carbonitride stringers that remain undissolved throughout the forging and heat-treating cycles. Even the double vacuum-melted alloys contain certain amounts of these constituents. Brittle carbides and nitrides do not elongate during forging. Rather, they form a discontinuous line of discrete particles. Increasing reductions, therefore, would be expected to improve transverse ductility in alloys containing these compounds. Carbonitrides, on the other hand, elongate during forging and usually promote increasingly anisotropic ductility with increasing reductions. Since vacuum melting removes most of the carbonitrides, vacuum-melted alloys exhibit a minimum level of mechanical-property directionality with respect to grain flow. Table 17-5 compares tensile properties of several parts forged from vacuum-melted superalloys which

TABLE 17-5. COMPARISON OF LONGITUDINAL AND TRANSVERSE MECHANICAL PROPERTIES OF SEVERAL SUPERALLOY FORGINGS[12]

Room-Temperature Properties of Die Forgings

Test Bar Location	Number of Tests	0.2% Offset Yield Strength, ksi Range	0.2% Offset Yield Strength, ksi Average	Tensile Strength, ksi Range	Tensile Strength, ksi Average	Elongation, per cent Range	Elongation, per cent Average	Reduction of Area, per cent Range	Reduction of Area, per cent Average
Part 1 - Waspaloy Compressor Hub									
Longitudinal	4	124.2-132.0	128	187.1-193.0	189	21.0-26.5	24.0	23.1-34.5	28.0
Transverse	24	120.8-136.0	127	174.0-198.4	184	15.4-25.0	22.2	19.1-28.5	21.0
Part 2 - Waspaloy Ring									
Radial	4	133.0-143.0	136	182.8-190.8	185	18.7-29.0	24.0	26.5-32.9	29.0
Rim tangential	6	132.0-137.0	135	177.6-185.2	180	21.8-28.1	25.0	26.5-35.1	31.0
Part 3 - René 41 Disk									
Rim tangential	2	142.3-143.8	143	193.5-201.7	197.6	14.0-18.0	16.0	16.0-18.2	17.1
Center tangential	2	138.9-142.1	140.6	196.0-187.0	191.0	18.0-20.0	19.0	20.3-23.6	21.9
Part 4 - Waspaloy Disk									
Rim tangential	3	130.7-135.0	132.7	189.9-191.9	190.5	23.6-26.1	25.4	22.7-30.4	26.4
Center tangential	2	127.5-135.4	131.5	185.0-188.6	186.8	24.0-25.0	24.5	25.3-27.3	26.3

exhibited rather small differences in mechanical properties for various test-bar locations.[12] Isotropic properties are reflected by the small variations in ductility with respect to test-bar locations representing longitudinal and transverse grain flow.

Grain Size. The final grain size of forged and heat-treated superalloys has an important influence on mechanical properties. With increasing grain size, tensile and stress-rupture ductility generally decrease and strain values become more erratic.

Dzugutov and Vakhtanov[13, 14] made a rather complete study of the influences of forging temperature, forging reduction, and grain size on the mechanical properties of a nickel-base superalloy similar to Nimonic 80A. Samples were forged with reductions varying between 4 and 80 per cent at several temperatures between 1560 and 2120 F. Grain sizes were measured on samples in both the as-forged and in the solution-heat-treated (1950 F, 8 hours, air cooled) conditions. Mechanical properties were determined after aging from either of these conditions (aging treatment: 16 hours at 1380 F). Table 17-6 shows the grain size of samples after 10, 25, and 50 per cent reductions at each of five forging temperatures. The study showed that the most uniform and finest grain size was obtained by forging the alloy at 1830 F. The widest variation in grain size was observed in samples forged at 2120 F.

Studies were also conducted on samples forged at 1900 F with reductions between 4 and 80 per cent to determine how much reduction was needed to remove all prior variations in grain size. After only 4 per cent reduction, the samples were recrystallized completely after solution heat treatment, but the grain-size variation remained

(ASTM 1-4). After about 12 per cent reduction, the solution-heat-treated samples were recrystallized to a uniform grain size of ASTM 4. In the as-forged condition, uniform grain size was obtained at reductions above 30 per cent. At 80 per cent reduction, the as-forged grain size was finer (ASTM 5-6) than after subsequent solution heat treatment (ASTM 4).

TABLE 17-8. INFLUENCE OF FORGING TEMPERATURE AND FORGING REDUCTION ON THE GRAIN SIZE OF A RUSSIAN ALLOY SIMILAR TO NIMONIC 80A[13,14]

Grain Size Equivalent to ASTM Standard Ratings

Heat Treatment Condition[(a)]	Forging Reduction, per cent	Grain Size of Alloy at Indicated Forging Temperature				
		1560 F	1650 F	1830 F	2020 F	2120 F
F	10	4	4	4	4	2-4
S	10	2-4	2-4	4	2-4	2-3
F	25	3-4	4	4	4	1-3
S	25	2-4	4	4	2-3	2-3
F	50	4	4	4	4	1-3
S	50	4	3-4	4	2-3	1-4

(a) F - As forged: samples air cooled from forging, then aged 16 hours at 1380 F.
S - Solution heat treated: samples solution annealed 8 hours at 1975 F, air cooled, then aged as above.

As expected, the best ductility for heat-treated samples was associated with the finest grain size; actual values were:

ASTM Grain Size	Room-Temperature Elongation, per cent
1-3	19-20
2-4	17
~4	24-26
4	26-29

On the other hand, the room-temperature strengths of forgings produced at temperatures ranging from 1560 to 2020 F were not noticeably influenced by grain size. However, samples forged at 2120 F exhibited 25 per cent lower ultimate strengths than those produced at lower temperatures. The loss of strength was probably associated more with the altered precipitation behavior than with grain-size effects.

To illustrate the importance of hot deformation on grain size, Figures 17-9 and 17-10 show etched sections from a disk and a blade forging, respectively, which exhibit unusually large grains. Both forgings received small reductions during the final forging operation. Both are shown after heat treatment. The Incoloy Alloy 901 disk was restruck at 2025 F and solution heat treated for 2 hours at 1985 F. The Inconel Alloy X-750 blade was coined at 2075 F then solution annealed for 2 hours at 2100 F. In both cases, the final reductions were in the vicinity of 5 to 10 per cent. Such small reductions are frequently responsible for promoting abnormal grain growth during heat treating as was the case for the two examples cited.

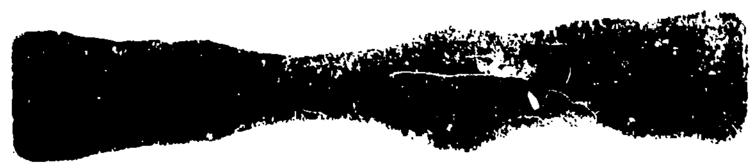


FIGURE 17-9. MACROSTRUCTURE OF FULLY HEAT TREATED DISK FORGING OF INCOLOY ALLOY 901 THAT WAS GIVEN LIGHT REDUCTION DURING FINAL FORGING STEP

One-half size

FIGURE 17-10. MACROSTRUCTURE OF FULLY HEAT TREATED BLADE FORGING OF INCONEL ALLOY X-750 THAT WAS GIVEN LIGHT REDUCTION DURING FINAL FORGING STEP

One-quarter size

The known methods for preventing abnormal grain growth in such parts are 1) to provide reductions greater than 15 per cent during the final operation and (2) to conduct restriking or coining operations at temperatures higher than the subsequent annealing temperatures. It is desirable to avoid restriking operations whenever possible. This is a factor to consider in forging design where forgings with more precise dimensional requirements usually require frequent restriking steps.

COMMENTARY ON FORGING PRACTICES

The forging of nickel-base superalloys requires close control over both metallurgical and operational conditions. Particular attention must be given to the control of the metal temperature on the die. Observers are usually required to record data on such items as transfer time, soaking time, finishing temperature, etc. Critical parts are frequently numbered to keep precise records of these factors on each forging. Records of dimensional variations are sometimes kept for each part to identify the levels of reduction obtained during forging. These steps are useful for identifying any forgings that contain material or forging defects, and permit metallurgical analysis so that the defects may be avoided in the future. In a sense, each forging and each design has characteristics that may or may not lead to satisfactory performance. Thus, such records can be used to avoid recurring problems and to help in optimizing the procedures for subsequent production runs.

The nickel-base superalloys are quite sensitive to minor variations in composition. Subtle variations can cause large variations in forgeability, grain size, and final properties. Zirconium has been shown to increase forgeability but, above very small quantities, it can promote cracking during forging and even during heating. One forging company experienced wide heat-to-heat variations in grain size in parts forged from Incoloy Alloy 901 in the same sets of dies. For some parts, it was found necessary to determine optimum forging temperatures for each incoming heat of material by making sample forgings and examining them after heat treatment for grain-size variations. Attempts to correlate composition with grain size were unsuccessful in this instance. The experience indicates that even subtle differences in production practices can be important.

Another company reported difficulty in maintaining uniform response to heat treatment in Waspaloy forgings until the effects of small variations in silicon, manganese, and titanium contents were established. Although the details of the practices were not disclosed, the response was improved by adjusting the time and temperature of the stabilizing heat treatment according to composition. Larger amounts of silicon appear to accelerate the precipitation reaction at the grain boundaries during this intermediate treatment, and stabilizing times are sometimes shortened to prevent excessive precipitation. It is reported that lower silicon and manganese contents lead to improved rupture ductility in Waspaloy forgings. No data were made available to substantiate this contention, however.

Companies experienced in the forging of nickel-base superalloys generally agree that the forging techniques developed for one part usually must be modified for producing another configuration from the same alloy. For this reason, development time often is needed to establish suitable forging and heat-treating cycles. This is especially true for the newer alloys (e.g., Waspaloy, Udimet 500, René 41, Udimet 700, etc.).

Heating for Forging. When heating nickel-base alloys, it is important to avoid contact with sulfur. Preferred fuels for forge furnaces are natural gas and low-sulfur oil. Furnaces used for carbon and low-alloy steels frequently contain sulfur-rich scale which should be removed before charging nickel alloys. A good practice consists of placing billets on clean bricks or plates of heat-resistant alloys. Furnace atmospheres are usually maintained slightly oxidizing.

Because of their lower thermal conductivity, nickel-base superalloys generally require longer soaking periods than steels. Longer times are also necessary to insure maximum solution of precipitated phases and compounds which dissolve slowly. Alloys having narrow working temperature ranges require more precise temperature control than ordinary steels. To achieve uniform billet temperatures, it is common practice to turn the billets and forgings while in the furnace. At least one company uses cast nickel-alloy billet cradles to elevate the billets above the furnace hearth. They are particularly useful when glasses are used for precoatings.

Volumetric changes occurring when heating the precipitation-hardenable alloys can cause internal cracking in billets and forgings containing large sections (3 inches or larger). Such cracks can go unnoticed until the forgings reach the final inspection step (usually ultrasonic). Practices successfully used to prevent these cracks include (1) preheating at temperatures corresponding to aging temperatures, soaking, then heating to forging (or heat treating) temperatures and (2) heating slowly (about 50 F/hr) between 1200 F and about 1600 F.

Ingot and Billet Breakdown. In the as-cast condition, nickel-superalloy ingots are generally characterized by coarse columnar grains that impair forgeability. Common mill practice consists of press forging between flat dies with small reductions with frequent reheatings at maximum forging temperatures. Billets are often ground to remove any cracks occurring during processing. After the cast grains are deformed, the billets usually exhibit better forgeability.

During fullering, drawing, rolling, cogging, or other types of side forging steps, nickel-base superalloys are sensitive to center bursts. Rapid forging strokes cause a "shear effect" at the center of the bar characterized by the appearance of an "x" at the ends of the bar. The "x" will be hotter than the rest of the bar and rupturing will occur in this region if deformations are too rapid or too large. By observing the relative brightness of the "x" pattern, experienced operators are able to control the reductions to avoid cracking. During upsetting the temperature increase is not as readily apparent, but center bursts are possible if forging is too rapid.

Lubrication. The choice of lubricants for forging nickel-base superalloys is limited to those free of sulfur. Lubricants containing molybdenum disulfide or other sulfur compounds are believed to have harmful effects on nickel alloys. The most common lubricants are mixtures of oil and graphite. Certain glasses are particularly useful for coating the alloys having narrow forging temperature ranges since they insulate as well as lubricate during forging. Because the nickel alloys resist oxidation, seizing and galling between the workpiece and dies are frequent problems. One company indicated that, if forging dies are tempered at about 750 F after machining, the oxide formed on the dies serves as an excellent parting agent and also helps to retain lubricant on the die surfaces.

Asbestos sheets, mica, and sawdust are sometimes used in conjunction with oil-graphite lubricants for forging superalloys. Such materials are used mainly for minimizing die chilling when forging parts with a minimum of detail and generous contours in hydraulic presses.

Control of Forging Reductions. The recrystallization temperatures of most superalloys are quite close to the forging temperatures. For example, René 41 recrystallizes in the vicinity of 1900 F. However, the process is apparently slow enough that the alloy exhibits work-hardening behavior even when forged at 2050 F. The knowledge of such behavior is important to avoid "critical-strain effects" and grain-growth problems. If the alloys are forged at temperatures and reductions equivalent to about 2 to 8 per cent cold reduction, coarse grains ordinarily develop during subsequent solution heat treatment. For this reason, it is important to conduct wedge-forging tests or other studies suitable for identifying the combinations of forging temperatures and reduction which lead to "critical-strain effects". Once the conditions are identified, they can be avoided by selecting suitable forging cycles. Two generally accepted forging principles are: (1) to finish all forging steps at true hot-working temperatures so that no recrystallization will occur during subsequent heat treatment or (2) to complete forging at hot-cold working temperatures with a reduction greater than about 15 per cent so that fine, recrystallized grains are obtained during subsequent solution heat treatments. The latter approach is favored for forgings containing thin rapid-cooling sections. For nickel-base superalloys, the maximum reduction is usually in the vicinity of 30 per cent.

For salvage forging operations (e.g., restriking forgings to correct dimensions, underfill, etc.) it is good practice to impart no less than 10 per cent reduction during forging. This sometimes requires a special die designed slightly smaller than the finishing die. Dies for salvaging disk shapes, for example, are usually designed with webs that are 10 to 15 per cent thinner than normal. Forgings that need restriking are often rough machined before forging to assure uniform reductions from part to part.

COMMENTARY ON FORGING DESIGN

The greatest use of nickel-base superalloy forgings is in turbojet-engine applications. For the most part, shapes required consist of disks, rings, and blades. From the previous discussions, it is apparent that the forging designs are restricted to comparatively simple, generously contoured shapes.* In the case of disks, for example, the webs are rarely thinner than 3/4 inch when forged from such alloys as René 41 or Udimet 500. As a rule, radius-to-thickness ratios fall between 5:1 and 15:1. For example, the normal web thicknesses for disks of three sizes would be as follows:

Disk Diameter, inches	Normal Web Thickness, inches
10	0.5-1.0
20	1.0-2.0
30	1.25-2.5

Thinner webs are possible by forging with additional reheating steps, but to do so increases the risk of "critical-strain effects".

*Forging-design practices for nickel-base superalloys are also described in Chapter 4.

Empirical forgeability ratings (Table 17-5) indicate the relative forging difficulty for various alloys, but these ratings are not directly relatable to specific forging designs or dimensions. Such ratings are useful in a general sense, however, since the more forgeable alloys can be forged in a greater variety of shapes and to smaller section sizes, and require the least number of forging operations.

It is helpful to look upon the nickel-base superalloys as consisting of two groups. The first group (Group I) would consist of the more readily forged alloys such as Incoloy Alloy 901, Hastelloy Alloy X, Inconel Alloy 718, and Nimonic 80A. The second group (Group II) includes alloys that are difficult to forge such as Udimet 500, Udimet 700, René 41, and Astroloy. Such groupings permit a more comprehensive discussion of forging designs.

Forging Size. Table 17-7 presents data on the size of typical forgings reported for several nickel-base superalloys. Representatives from forging companies report that the size potential for the superalloys is in the neighborhood of 1800 plan inches for simple shapes and probably half that for shapes having ribs and webs. Limits on section size result from forging-pressure requirements. Small forgings, such as blades, are rarely forged to section thicknesses less than 1/4 inch from the Group II alloys. In hammer forging, there is always the risk of cracking caused either by excessive temperature increases or strain-rate effects. Practical section thicknesses are given in Table 17-8 for various sizes of die forgings from alloys in each of the two groups.

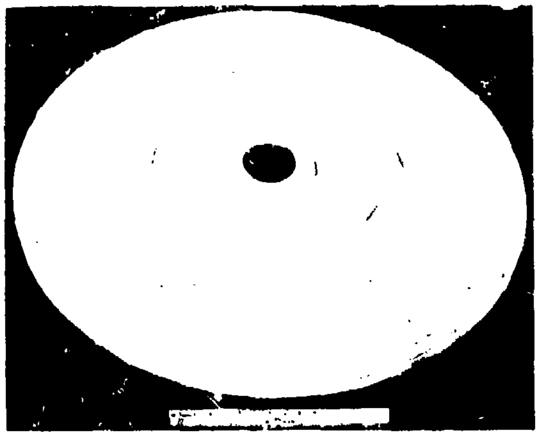
Availability. Figure 17-11 illustrates the most common designs that have been produced successfully from nickel-base superalloys in production-lot quantities. The part illustrated in Figure 17-12 is typical of the type of structural designs that can be forged from the Group II alloys. The impeller forging shown in Figure 17-13 was made on an experimental basis from Waspaloy and on a semi-production basis from Inconel Alloy X-750. This intricate design is not considered practical because of high tooling and processing costs. However, the part does illustrate that such close-tolerance designs are within the technical capability of present-day forging technology.

TABLE 17-7. SIZES OF FORGINGS SUCCESSFULLY FORGED FROM SEVERAL NICKEL-BASE SUPERALLOYS

Alloy	Forging-Difficulty Group	Type of Forging	Forging Size: Maximum Dimension, in.	Forging Size: Weight, lb
Incoloy Alloy 901	I	Hub	22 long	190
		Disk	45 dia.	650
Inconel Alloy X-750	I	Impeller	8 dia.	--
René 41	II	Disk	26 dia.	241
		Disk	45 dia.	--
		Hub	20 dia.	200
		Disk	44 dia.	950
		Disk	39 dia.	950
Udimet 500	II	Disk	45 dia.	700
Waspaloy	II	Disk	34 dia.	550
		Disk	33 dia.	250
		Disk	17 dia.	180
Astroloy	II	Disk	36 dia.	170

TABLE 17-8. SUGGESTED MINIMUM SECTION SIZES FOR NICKEL-BASE SUPERALLOY FORGINGS

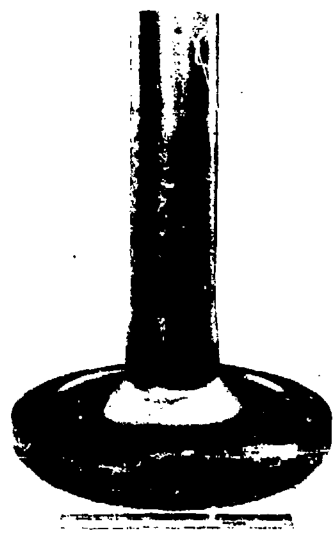
	Minimum Section Size, inch: Group I	Minimum Section Size, inch: Group II
Small Forgings, up to 10 lb	3/16	3/8
Medium Forgings, 10 to 50 lb	1/2	3/4
Large Forgings, over 50 lb	3/4	1



A. Udimet 500 Turbine Disk
Diameter: 45 inches

B. Astroloy Turbine Disk
Diameter: 26 inches

C. René 41 Conical Shaft
Diameter: 20 inches



D. Incoloy Alloy 901 Shaft and Disk
Disk Diameter: ~15 inches
Shaft Length: ~18 inches

FIGURE 17-11. TYPICAL NICKEL-BASE SUPERALLOY PARTS FORGED TO BLOCKER AND COMMERCIAL DESIGNS IN PRODUCTION

A, B, and C – Courtesy Wyman-Gordon Company
D – Courtesy Cameron Iron Works, Inc.

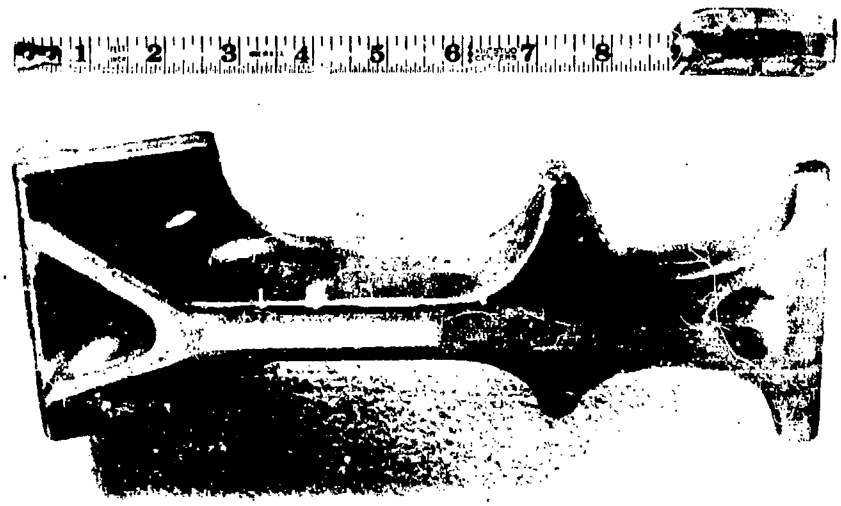


FIGURE 17-12. TYPICAL STRUCTURAL FORGING DESIGN PRODUCIBLE FROM NICKEL-BASE SUPERALLOYS ON A COMMERCIAL BASIS

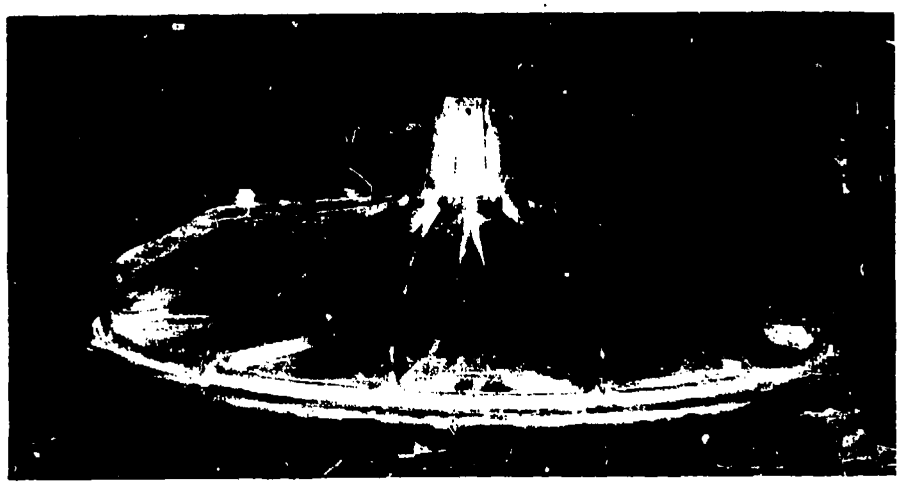


FIGURE 17-13. WASPALOY IMPELLER IS AN EXAMPLE OF CLOSE-TOLERANCE FORGING DESIGN PRODUCIBLE FROM NICKEL-BASE SUPERALLOYS ON AN EXPERIMENTAL BASIS

REFERENCES

(1) Weisenberg, L. A., and Morris, R. J., "Manufacture of René 41 Components" Metal Progress, 78, 70-74 (November, 1960).

(2) Moon, D. P., Baker, J. F., Simmons, W. F., "Effect of Elevated Temperature Exposure on the Strength Properties and Microstructure of René 41 and Astroloy", Battelle Memorial Institute and General Electric Company, Cincinnati, Ohio, Paper presented at the 42nd Annual Meeting of ASM, Philadelphia (October 17-21. 1960).

(3) Decker, R. F., Rowe, J. P., and Freeman, J. W., "Boron and Zirconium from Crucible Refractories in a Complex Heat-Resistant Alloy", NACA Report 1392 (1958).

(4) Kostock, F. R., "Material Properties Data Sheets", North American Aviation, Inc., Los Angeles, California (March 15, 1961).

(5) Couts, Jr., W. H., "Program to Develop a Wrought Blade Superalloy", General Electric Company, Cincinnati, Ohio, APEX, for USAF and AEC, Progress Report on Contract AF 33(600)-38062 and AT (11-1)-171 (September 1, 1961).

(6) Alloy Performance Data Brochures, Metals Division, Kelsey-Hayes Company, New Hartford, New York.

(7) Hastelloy Alloy Brochures, Union Carbide Stellite Company, Division of Union Carbide Corporation, Kokomo, Indiana.

(8) Private Communicatior from E. A. Greer of Lockheed Aircraft Corporation, Burbank, California.

(9) Nichols, H. R., Graft, W. H., et al., "Development of High Temperature Die Materials", Final Engineering Report, Contract No. AF 33(600)-35915, Armour Research (October, 1959).

(10) Metals Handbook, Vol. 1: Properties and Selection of Metals, Eighth Edition, American Society of Metals, Cleveland, Ohio (1961).

(11) Nippes, E. F., Swager, W. F., Grotte, G., "Further Studies On the Hot Ductility of High-Temperature Alloys", Report No. 33, Bulletin Seven, Welding Research Council of the Engineering Foundation, New York, New York (February, 1957).

(12) Henning, H. J., and Boulger, F. W., "Superalloy Forgings", DMIC Memorandum No. 86, Defense Metals Information Center, Battelle Memorial Institute (February 10, 1961).

(13) Dzugutov, M. Ya., and Vakhtanov, B. F., "Influence of the Deformation Conditions and the Subsequent Thermal Treatment on the Properties of Alloy EI437B", Kuznechno-Shtampovochnoe Proizvodstvo, No. 3, 3-7 (1961).

(14) Dzugutov, M. Ya., and Vakhtanov, B. F., "Influence of the Original Grain Size of an Alloy of EI437B Type on the Final Grain Size Obtained After Deformation and Recrystallization", Kuznechno-Shtampovochnoe Proizvodstvo, No. 2, 1-4 (1960).

CHAPTER 18

COLUMBIUM AND COLUMBIUM ALLOYS

Crystalline Structure:	Body-centered cubic
Phase Changes:	None
Number of Phases:	One at forging temperatures
Solidus Temperature:	Unalloyed – 4475 F Alloys – 4100 to 4700 F
Reaction When Heated in Air:	Continuous oxidation above about 1000 F; gas contamination above about 750 F

ALLOYS AND FORMS AVAILABLE

Nominal compositions and forging characteristics of columbium and several columbium alloys are given in Table 18-1. Relatively little forging, other than in ingot breakdown operations, has been done on columbium alloys; most of the closed-die forging experience has been with unalloyed columbium and the Cb-1Zr alloy. Hence, the forging characteristics indicated for several alloys, by necessity, are based on information available for breakdown fabrication of cast ingot to sheet bar and heavy plate.

Columbium and several of its alloys (such as Cb-1Zr and Cb-33Ta-1Zr) are readily cold workable and may be forged directly from as-cast ingot. The more highly alloyed compositions containing tungsten or molybdenum (such as Cb-15W-5Mo-1Zr) generally require hot breakdown fabrication by extrusion to overcome undesirable features of the cast structure and insure adequate forgeability. After initial breakdown, most of these alloys are ductile below room temperature. This has been evidenced by the ability to roll many of the alloys with up to 90 per cent reduction at or near room temperature.

Columbium and its alloys are generally prepared commercially in the form of cylindrical cast ingots by consumable-electrode, vacuum-arc or electron-beam melting. The most common ingot sizes are 6 to 8 inches in diameter, but ingots up to 16 inches in diameter are reported to be available for some alloys.

TABLE 18-1. NOMINAL COMPOSITION AND FORGING CHARACTERISTICS OF COLUMBIUM AND SEVERAL COLUMBIUM ALLOYS[1,2,3]

Nominal Composition, w. %	Designation	Approx. Solidus, F	Approx. Minimum Recrystallization Temperature, F	Approx. Minimum Hot Working Temperature, F[a]	Forging Temperature, F Normal Range	Lowest Reported	Forgeability
99.2 Cb	Unalloyed	4475	1900	1500	RT-2000	RT	Excellent
Cb-1Zr	FS-80	4350	1900	2100	RT-2300	RT	Excellent
Cb-33Ta-1Zr	FS-82	4570	2200	2400	1900-2700	RT	Good
Cb-28Ta-10W-1Zr	FS-85	4695	2250	2400	2300-2500[c]	--	Good[c]
Cb-10Ti-10Mo-0.1 C	D-31	4100	2200	2500	1900-2700	1900	Moderate
Cb-10W-1Zr-0.1 C	D-43	4700	2100	2200	2000-2200[c]	--	Moderate[c]
Cb-10W-5Zr	Cb-74[b]	4400	2200	2500	2200-2600	2200	Moderate
Cb-10W-2.5 Zr	Cb-752	--	2100	2300	2200-2600[c]	--	Good[c]
Cb-15W-5Mo-1Zr	F-48	4500	2600	3000	2400-3000	1850	Fair
Cb-10Ta-10W	SCb-291	4710	2100	2400	1700-2200	1600	Good
Cb-5V-5Mo-1Zr	B-66	4300	2100	2400	2200-3000[c]	--	Moderate[c]
Cb-10W-10Hf-0.1Y	C-129Y	--	2000	2200	2000-2200[c]	--	Good[c]

(a) Minimum hot working temperature is defined as the lowest temperature at which the alloy will begin to recrystallize during forging. Below these temperatures, the alloys are cold worked.
(b) Production of alloy discontinued; replaced by Cb 752.
(c) Based on ingot breakdown and breakdown rolling experience.

FORGING BEHAVIOR

Alloys containing titanium, vanadium, tantalum, molybdenum, and tungsten are solid-solution strengthened by these additions. These alloying elements also raise the recrystallization temperature and enhance the retention of elevated temperature strength. Zirconium and hafnium dispersion harden columbium by reacting with oxygen or carbon present in the base materials as a residual impurity or as an intentional addition. Alloys containing large amounts of dispersed phases generally have reduced forgeability. Since most of the alloys have been designed for high strength at elevated temperature, they commonly contain both types of alloying elements.

Primary working (i.e., cast ingot breakdown) of the high-strength alloys requires temperatures from about 2200 F to 3000 F. Secondary working operations such as closed-die forging and rolling can then be carried out at temperatures from about 2500 F to as low as 1000 F. Because most of the alloys in Table 18-1 can withstand large amounts of hot-cold work without intermediate annealing, once in the wrought condition, it should be possible to select forging temperatures that represent a best compromise between forgeability and forging-pressure requirements.

Columbium and its alloys have great affinity for oxygen at elevated temperatures. Exposure to air at hot-working temperatures not only forms a hard, brittle surface oxide, but also an oxygen contamination zone within the metal, which progresses inward with continued heating. Oxygen contamination not only has an adverse effect on properties, but can seriously impair forgeability. To avoid oxygen embrittlement, the surface of the workpiece must be protected during heating and, if practicable, during the hot-working operation.

Darby and Carson[4] investigated the upset forgeability of four likely columbium forging alloys after breakdown of the arc-cast ingots by extrusion. The various conditions under which the alloys were upset and the results of the forging trials are summarized in Table 18-2. The adverse effects of oxygen contamination on forgeability were evidenced in these studies by the occurrence of cracking of all of the alloys wherever the protective coatings failed. With suitable protection, the applicability of the alloys for closed-die forging, based first on forgeability and then on forging pressure, was rated in the following order: Cb-10Ti-10Mo-0.1C, Cb-10W-5Zr, Cb-15W-5Mo-1Zr, and Cb-20W-10Ti-6Mo.

TABLE 18-2. SUMMARY OF FORGEABILITY STUDIES ON FOUR ARC-CAST AND EXTRUDED COLUMBIUM ALLOYS[4]

Sample size: 1-1/2-in. diam x 3 in. long
Coating: Al-10Cr-2Si
Heating atmosphere: Still air

Alloy	Cb-10Ti-10Mo-0.1C	Cb-20W-10Ti-6Mo	Cb-10W-5Zr	Cb-15W-5Mo-1Zr
Forging Temperature	2000, 2200, 2400 F	2300, 2500, 2700 F	2200, 2400, 2600 F	2300, 2500, 2700 F
Upset Reduction	(a) 50% in one heating (b) 75% (50% in one heating, another 50% after reheating)			
Forgeability	Successfully upset 75% without cracking; some samples had slight checking on flats.	Successfully upset 50% at 2500 and 2700 F; slight cracking at 2300 F. Could not achieve 75% without cracking.	Successfully upset 75% without cracking; slight checking on flats.	Successfully upset 50% at all temperatures. Upset 75% without cracking at 2700 F.
No. of Hammer Blows Required for 50% Upset	9 to 12 blows	12 to 13 blows	33 to 39 blows at 2200 F 12 to 14 blows at 2400-2600 F	10 to 12 blows

Similar forging behavior was reported by Nemy[5] in upsetting studies on the Cb-10Ti-10Mo-0.1C and Cb-20W-10Ti-6Mo alloys. The Cb-10Ti-10Mo-0.1C alloy exhibited much higher lateral ductility and required considerably lower pressures. Figure 18-1 shows the pressure and temperature requirements for upsetting the two alloys to a 25 per cent reduction. A similar curve for Udimet 500 is included for comparison, and illustrates the relatively wide range of forging temperatures common to most columbium alloys.

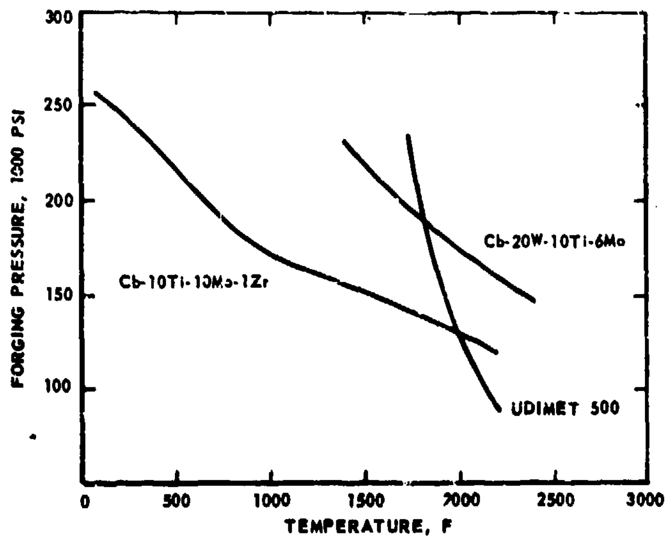


FIGURE 18-1. FORGING PRESSURES OF TWO COLUMBIUM ALLOYS FOR 25% UPSET REDUCTION AT VARIOUS TEMPERATURES

DATA FOR UDIMET 500 SHOWN FOR COMPARISON.

AFTER NEMY[5]

In forging jet-engine blades from these two columbium alloys at Thompson-Ramo-Wooldridge[5], two different procedures for producing the blade preform were required because of the differences in forgeability. The Cb-10Ti-10Mo-0.1C alloy preforms could be produced by a combination of end upsetting and roll forging. Both of these operations require good lateral ductility. With the Cb-20W-10Ti-6Mo alloy, the preforms had to be made from larger bar stock by a combination of extrusion and a low-reduction, closed-die blocker forging operation.

Forging pressures for the various columbium alloys can be estimated to some extent from the elevated-temperature properties given in Table 18-3. Included in the table are data for unalloyed molybdenum and for the Mo-0.5Ti-0.08Zr (TZM) alloy to permit comparisons. These data indicate that some of the columbium alloys are much stronger than the Mo-0.5Ti-0.08Zr (TZM) alloy. For example, at 2000 F the Cb-15W-5Mo-1Zr alloy is about twice as strong as the molybdenum alloy.

Most likely, columbium-alloy forgings will be used for many future applications in the wrought and stress-relieved condition to take advantage of the higher strengths that are possible by hot-cold working. Ingram and Ogden[6] have summarized the general behavior of refractory metals and alloys under hot-cold working conditions that would be representative of many closed-die forging operations. Although few data are presented for columbium and its alloys, they provide some insight to forging behavior under hot-cold working conditions and the effects of forging variables. In the hot-cold working range, the mechanical properties of columbium and columbium alloys will be strongly influenced by:

(1) Amount of reduction after the final anneal
(2) Working temperature
(3) Intermediate annealing temperature.

Similarly, these variables will influence forging behavior and, to a large extent, govern the actual forging sequence and procedures.

TABLE 18-3. COMPARATIVE TENSILE PROPERTIES OF COLUMBIUM AND SEVERAL COLUMBIUM ALLOYS AT TEMPERATURES CORRESPONDING TO TYPICAL FORGING TEMPERATURES[1,2,3]

Data for Molybdenum and TZM Included for Comparison

Alloy	Temperature, F	Tensile Strength, 1000 psi	Yield Strength, 1000 psi	Elongation, per cent
Unalloyed Cb*	RT	40.0	30.0	30
	1800	12.5	--	37
	2000	10.0	8.0	42
Cb-1Zr	RT	48.0	35.5	15
	2000	23.0	20.5	14
	2500	8.5	6.5	>50
Cb-28Ta-10W-1Zr	2000	36.0	29.5	27
	2300	25.0	22.0	36
	2600	14.5	14.0	79
Cb-10Ti-10Mo-0.1C	2000	29.0	25.5	36
	2300	17.5	--	86
	2600	10.0	8.0	100
Cb-10W-2.5Zr	2000	[illegible]	31.5	18
	2200	33.5	26.0	27
	2400	23.0	18.5	44
Cb-15W-5Mo-1Zr	2000	64.0	58.0	22
	2300	50.0	46.0	30
	2600	32.0	31.0	56
Unalloyed Mo	2200	21.0	11.0	38
	2400	11.0	6.0	57
Mo-0.5Ti-0.08Zr (TZM)	2000	35.5	21.0	[illegible]
	2500	26.0	16.5	19
	3000	17.5	11.0	20

In working the alloys at temperatures below their recrystallization temperatures, higher forging pressures will be required with progressively increasing amounts of deformation. This is illustrated in Figure 18-2 by the room-temperature yield strength of columbium and two of its alloys after various amounts of cold work. It can be seen that the rate of work hardening is greater for the alloys than for unalloyed columbium, and that the yield strength can be increased as much as 25,000 psi by reductions in the order of 50%. Increasing the amount of cold reduction, however, also lowers the recrystallization temperature.

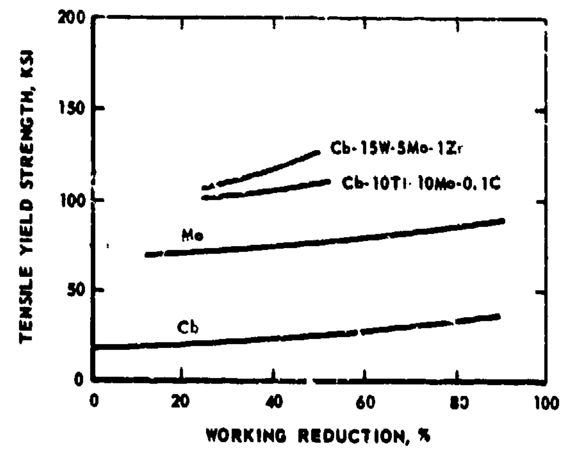


FIGURE 18-2. EFFECT OF COLD WORKING REDUCTION ON THE STRENGTH OF COLUMBIUM AND COLUMBIUM ALLOYS AT ROOM TEMPERATURE(6)

DATA FOR MOLYBDENUM INCLUDED FOR COMPARISON

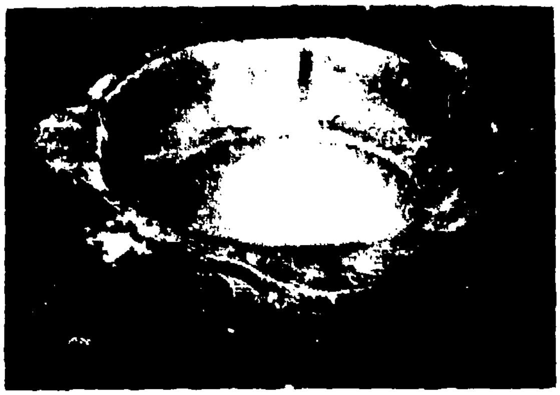
Varying the working temperature below the minimum recrystallization temperature does not have significant effects on strength, since strength is controlled primarily by the amount of strain. However, in the hot-cold working region, the working temperature exerts a pronounced effect on the amount of strain required to reach a desired strength level. As the working temperature is increased, the working reduction must also be increased to result in a corresponding amount of strain hardening.

COMMENTARY ON FORGING PRACTICES

Alloys such as Cb-1Zr and Cb-33Ta-1Zr are readily forgeable directly from as-cast ingot. Two examples of large Cb-1Zr die forgings, weighing up to 1100 pounds, produced directly from arc-cast ingots are shown in Figure 18-3. Die forging of the Cb-1Zr alloy is quite similar to that for alpha-titanium alloys, except that greater precaution must be taken against contamination and higher forging pressures are required.

As shown in Figure 18-3, parts directly forged from cast ingots may contain shallow surface laps and wrinkles traceable to the coarse grain structure. For such large forgings, therefore, ample allowances must be made for surface finishing to compensate for both wrinkling and contamination. Alternatively, to avoid excessive material losses, it may be desirable to initially forge the ingot between flat dies to break up the coarse grain structure before finish forging.

In their studies on the Cb-10W-5Zr alloy, Darby and Carson(4) reported that extruded and recrystallized material has adequate workability and is amenable to closed-die forging. Examples of a small part hammer forged at 2200 to 2400 F in this alloy are



Size: 35-in.-diam x 13-in. high

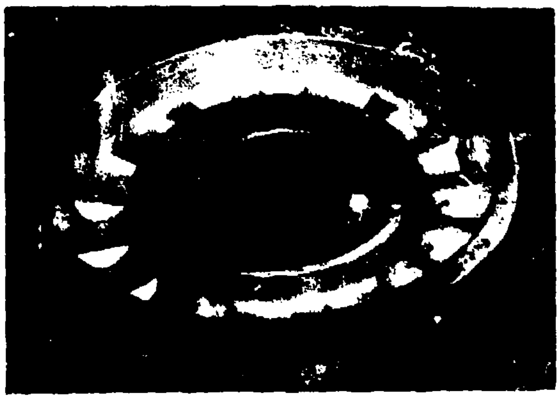
Weight: approx. 1100 lb.

Draft: 3°

Min. walls: 0.47 in.

Fillets: 1-in. R

Corners: 0.25-in. R



Size: 37-in.-diam x 6-1/2-in. high

Weight: approx. 850 lb.

Draft: 7°

Min. web: 0.83 in.

Min. rib: 1.5 in.

Fillets: 1-in. R

Corners: 0.25-in. R

FIGURE 18-3. Cb-1Zr ALLOY PARTS FORGED FROM AS-CAST INGOTS AT 2250 F

Courtesy of Wyman-Gordon Company

shown in Figure 18-4. Some difficulties were experienced in achieving complete die fill because of a pronounced tendency of the alloy to flow laterally, as evidenced by the large amount of flash on the parts. Satisfactory protection against atmospheric contamination of these parts was achieved with an Al-10Cr-2Si alloy coating applied by dipping at 1500 F and subsequently diffusing at 1900 F in argon. This coating when applied in thicknesses of 2 to 4 mils was found particularly useful between 2000 and 2600 F.

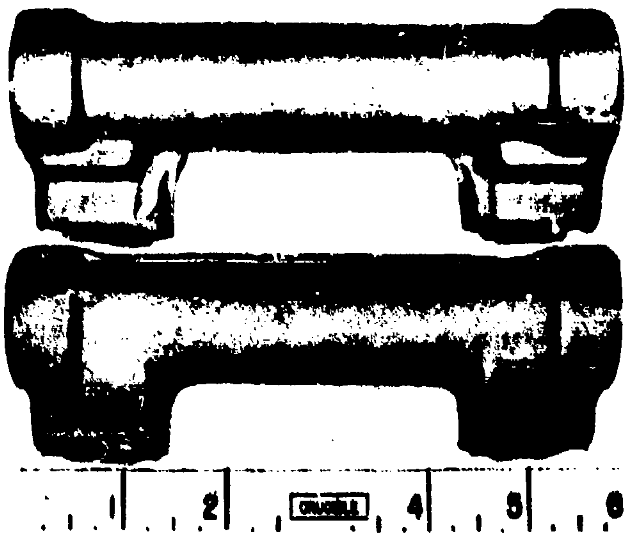


FIGURE 18-4. Cb-10W-5Zr CLOSED-DIE FORGINGS PRODUCED AT 2200 TO 2400 F(4) (AS-FORGED, FLASH REMOVED)

Draft: 7°

Corners: 1/16-in. R

Fillets: 1/8-in. R

Tolerances: +3/64 - 1/64 in.

Courtesy of Crucible Steel Company of America

The forging-pressure requirements for alloys such as Cb-10W-5Zr are considerably higher than normally encountered in routine production forging operations. As a rough comparison, Darby[4] estimated that, on a given part of 8640 steel that could be forged on a 2000-pound hammer and in titanium on a 5000-pound hammer, the columbium alloy would probably require a 14,000-pound hammer. Accordingly, the combination of high temperatures and high pressures required to forge such columbium alloys would dictate that stronger and more wear-resistant, high-temperature die materials would be required than normally used in routine production forging operations.

A typical shop forging practice is illustrated by the procedure for making large rings of the Cb-10Ta-10W alloy described by Perryman[7]. The finished dimensions of the ring, contoured on both the inside and outside faces, were 35-1/2 in. OD by 30 in. ID with a 3-3/8 in. face height (Figure 18-5). Starting with 17-1/2-inch long sections cut from conditioned 8-inch-diameter ingots and weighing about 300 pounds, the rings were produced by the following sequence of operations:

(1) Upset ingot sections to pancakes approximately 5-1/2 to 6 inches high by 14 to 15 inches in diameter on a 1500-ton forging press.

(2) Punch 5-3/4-inch-diameter hole in pancake to form heavy-wall ring.

(3) Mandrel forge to a ring having a 10 to 11 inch ID and a 3 to 4 inch wall thickness on a 3500-pound steam hammer.

(4) Cool and condition surface by spot grinding.

(5) Straight roll on a ring rolling mill to a ring approximately 25 in. OD by 1[illegible] in. ID, flatten on hammer.

(6) Contour roll on ring mill to finish dimensions.

FIGURE 10-5 CONFIGURATION OF CONTOURED RINGS PRODUCED FROM Cb-10Ti-10W ALLOY[7]

The temperatures and heating cycles for each operation were as follows:

Operation	Working Temperature, F Maximum	Minimum	Number of Reheats
Upset and punch	2100	1750	2
Mandrel forge	2100	1600	1
Straight roll	2100	1800	4
Contour roll	2100	1700	9

Starting billets were coated with a glass frit and heated in a gas-fired furnace with a slightly reducing atmosphere. The over-all oxidation loss, based on the difference between the starting billet weight and the as-rolled ring weight, averaged about 10 per cent of the starting billet weight. Surface removal to a depth of 1/8 inch by rough machining eliminated all effects of surface contamination. After vacuum annealing, the rings exhibited a recrystallized grain size of ASTM 4 to 6, and mechanical properties comparable to those obtained in sheet.

REFERENCES

(1) Bacon, F. E., et al., "The Processing and Properties of Columbium-Base Sheet and Forging Alloys – A State-of-the-Art Survey", Progress Report on Contract No. AF 33(600)-39942, Crucible Steel Company of America, December 2, 1959.

(2) Bartlett, E. S., and Houck, J. A., "Physical and Mechanical Properties of Columbium and Columbium-Base Alloys", DMIC Report 125, Defense Metals Information Center, Battelle Memorial Institute, February 22, 1960.

(3) Schmidt, F. F., and Ogden, H. R., "The Engineering Properties of Columbium and Columbium Alloys", DMIC Report 188, Defense Metals Information Center, Battelle Memorial Institute, September 6, 1963.

(4) Darby, P. F., and Carlson, R. O., "The Development of Optimum Manufacturing Methods for Columbium Alloy Forgings", Technical Documentary Report No. ASD-TDR-63-110, Contract No. AF 33(600)-39944, Crucible Steel Company of America, December 1962.

(5) Nemy, A. S., "Forging of the Refractory Metals", High Temperature Materials II, pp 561-575, Interscience, New York (1963).

(6) Imgram, A. G., and Ogden, H. R., "The Effect of Fabrication History and Microstructure on the Mechanical Properties of Refractory Metals and Alloys", DMIC Report 186, Defense Metals Information Center, Battelle Memorial Institute, July 10, 1963.

(7) Perryman, J. T., "The Forming of the SCb-291 Alloy", paper presented at the Technical Conference on Applied Aspects of Refractory Metals, sponsored by the AIMME, Los Angeles, California, December 9-10, 1963.

CHAPTER 19

TANTALUM AND TANTALUM ALLOYS

Crystalline Structure:	Body-centered cubic
Phase Changes:	None
Number of Phases:	Usually one at forging temperature
Solidus Temperature:	Unalloyed – 5425 F Alloys – 4400 to 5500 F
Reactions When Heated in Air:	Continuous oxidation above 1100 F; gas contamination above about 700 F

ALLOYS AND FORMS AVAILABLE

Table 19-1 gives the nominal compositions and forging characteristics of tantalum and several tantalum alloys that have been successfully forged.(1,2) Applications for most of these compositions have been primarily in the form of sheet; thus, most of the forging experience has been concerned with the conversion of ingots to sheet bar. Die and ring forgings have been successfully produced on a limited scale from unalloyed tantalum and from the Ta-10W alloy.

TABLE 19-1. NOMINAL COMPOSITION AND FORGING CHARACTERISTICS OF TANTALUM AND SEVERAL TANTALUM ALLOYS(1,2)

Nominal Composition, wt%	Approx. Solidus, F	Approx. Minimum Recrystallization Temperature, F	Approx. Minimum Hot Working Temperature, F[a]	Normal Forging Temperature Range, F[b]	Forgeability[b]
99.9Ta	5425	2000	2400	RT-2000	Excellent
Ta-10W	5495	2400	3000	1800-2300	Good
Ta-12.5W	5520	2750	>3000	>2000	Good
Ta-30Cb-7.5V	4400	2200	2800	2200-2400	Good
Ta-8W-2Hf	5400	2800	>3000	>2000	Good
Ta-10W-5Hf	5420	2400	3000	2100-2300	Fair

(a) Minimum hot working temperature is defined as the lowest forging temperature at which the alloys begin to recrystallize during forging. Below these temperatures, the alloys are cold-worked.

(b) Based on ingot breakdown and breakdown rolling experience.

Unalloyed tantalum and its alloys are prepared commercially by powder metallurgy, consumable-electrode arc melting, and electron-beam melting. Billets prepared by powder-metallurgy techniques are not amenable to direct forging, but most of the alloys listed in Table 19-1 can be directly forged from cast ingots. The most common ingot sizes are 6 to 8 inches in diameter but ingots up to about 16 inches in diameter are reported to be available for some alloys.

FORGING BEHAVIOR

Unalloyed tantalum and most of the single-phase alloys listed in Table 19-1 are readily forgeable directly from cast ingot. To reduce the resistance to deformation and to overcome undersirable effects of the coarse-grained cast structure such as wrinkles, laps, and internal cracking, initial breakdown forging operations are almost always performed at fairly high temperatures in the hot-cold working range – generally between 2000 and 2400 F. After about 50 per cent reduction, the forging temperatures for several of the alloys can be reduced below 2000 F with no difficulty.

The forging behavior of the Ta-10W alloy, for which there has been the most shop production experience, is representative of most of the single-phase tantalum alloys. Interstitial impurities such as carbon, oxygen, and nitrogen have a deleterious effect on forgeability. Torti[3] reported that the forgeability of the Ta-10W alloy is drastically reduced as the carbon level increases from 15 to 120 ppm. This is illustrated in Figure 19-1 by the billets upset at 2000 F from small ingots with carbon levels in this range. Similarly, oxygen at levels much above 100 ppm causes difficulty in initial breakdown and in conditioning the ingots. To permit direct forging of cast alloy ingots, the carbon, oxygen, and nitrogen levels are each held to less than 50-75 ppm, and the total interstitial content to less than 150 ppm.

FIGURE 19-1. FORGEABILITY TEST SAMPLES OF Ta-10W UPSET FORGED AT 2000 F FROM SMALL INGOTS(3)

	Analysis, ppm		
	C	O	N
Top	120	10	25
Center	60	24	29
Bottom	15	35	25

Courtesy of National Research Corporation

Ta-10W alloy ingots can be readily forged at 2000-2200 F. According to National Research Corporation[3], breakdown forging below 1800 F, or continued working below 1500 F can cause internal cracking. Binary Ta-W alloys with higher tungsten contents also can be directly forged, but higher temperatures are required than for the Ta-10W alloy and yields are poorer. Several forging vendors have observed that the forgeability

of Ta-W alloys at 2000-2200 F decreases sharply (i.e., conditioning losses increase sharply) at tungsten levels greater than 12.5 per cent. For this reason, 12.5 per cent tungsten is regarded as the maximum permissible in tantalum ingot for adequate forgeability in direct forging.

In general, the single-phase, or solid-solution, alloys of tantalum are more forgeable than the two-phase alloys. For example, Battelle[4] observed that the single-phase Ta-30Cb-7.5V alloy exhibited much better forgeability than the two-phase Ta-10Hf-5W alloy. The Ta-10Hf-5W alloy contains a hafnium-rich phase in the tantalum-rich matrix as an equilibrium constituent at temperatures below 2700 F. Thus, minimum requirements for direct forging this alloy would probably involve the use of a homogenization treatment and ingot preheating temperatures well above 2700 F. Through the use of extrusion at temperatures above 3000 F to break down the cast ingot structure, however, the Ta-10Hf-5W alloy can be forged readily at temperatures as low as 2150 F. By comparison, Battelle's experience with the Ta-30Cb-7.5V alloy indicates that direct forging of cast ingots at temperatures of about 2400 F should be practicable.

Like columbium, tantalum, and its alloys are contaminated by exposure to air at elevated temperatures. Oxidation at forging temperatures not only forms a hard, brittle surface oxide, but also an oxygen contamination zone below the surface, which progresses inward with continued heating. This is usually not detrimental to workability in heavy sections, but can seriously impair forgeability where repeated heatings would be necessary to die forge a part with thin sections.

For lack of actual forging pressure or flow stress data, forging pressures for the various tantalum alloys can be estimated to some extent from the elevated-temperature properties listed in Table 19-2. As these data show, the Ta-10W, Ta-12.5W and Ta-10Hf-5W alloys have about the highest yield strengths over their normal forging temperature ranges. Since the alloys are generally forged at hot-cold working temperatures, the forging pressures will increase with progressively higher reductions. This is illustrated in Figure 19-2 by the increase in room-temperature yield strength of tantalum and the Ta-10W alloy with increasing amounts of cold work[5]. Limited forging data for the Ta-30Cb-7.5V alloy upset between flat dies in a hydraulic press at 2200 F show that actual forging pressures increased from about 40,000 psi at the start of deformation to about 130,000 psi at 50 per cent reduction[4].

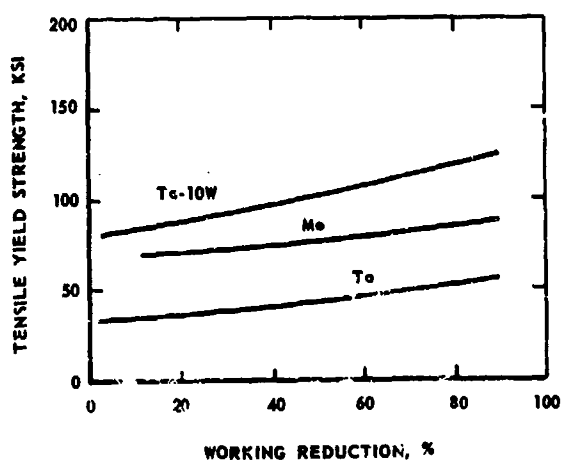


FIGURE 19-2. EFFECT OF COLD WORKING ON THE STRENGTH OF TANTALUM AND Ta-10W ALLOY AT ROOM TEMPERATURE[5]

DATA FOR MOLYBDENUM INCLUDED FOR COMPARISON

TABLE 19-2. COMPARATIVE TENSILE PROPERTIES OF TANTALUM AND SEVERAL TANTALUM ALLOYS AT TEMPERATURES CORRESPONDING TO TYPICAL FORGING TEMPERATURES[1, 2]

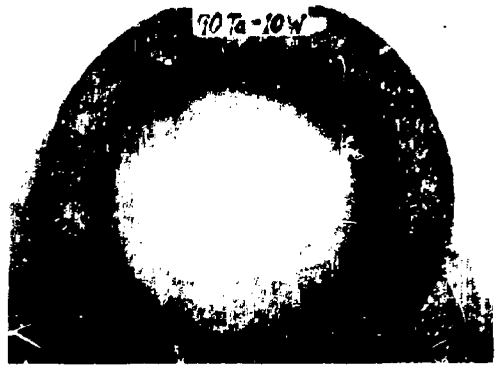
Data for Molybdenum and TZM Included for Comparison

Alloy	Temperature, F	Tensile Strength, 1000 psi	Yield Strength, 1000 psi	Elongation, per cent
Unalloyed Ta	RT	67.0	57.5	25
	1600	22.0	12.0	33
	2000	17.0	8.0	43
Ta-10W	2000	60.5	55.0	11
	2400	39.5	34.5	19
	3000	19.0	16.5	84
Ta-12.5W	2200	50.5	38.0	18
	2700	33.0	23.0	30
	3000	22.0	16.5	55
Ta-30Cb-7.5V	1800	76.0	53.0	38
	2200	46.0	33.5	51
	2400	33.0	27.0	87
	2700	19.0	16.0	>100
Ta-8W-2Hf	2000	61.0	35.0	18
	2200	49.5	28.5	25
	2700	31.0	24.5	30
	3000	15.0	12.0	48
Ta-10Hf-5W	2200	50.0	39.0	8
	2400	41.5	30.5	10
	2600	37.0	25.0	17
	3000	18.5	15.0	44
Unalloyed Mo	2200	21.0	11.0	38
	2400	11.0	6.0	57
Mo-0.5Ti-0.08Zr (TZM)	2000	35.5	21.0	20
	2500	26.0	16.5	19
	3000	17.5	11.0	20

a. Upset Forging

b. Punching

c. Mandrel Forging



d. Finished Ring Forging

FIGURE 19-3. SEQUENCES IN FORGING RINGS OF Ta-10W ALLOY AT 2000 TO 2200 F

Courtesy of Coulter Steel and Forge Company

COMMENTARY ON FORGING PRACTICES

Most shop forging experiences with tantalum alloys have been on the Ta-10W alloy and, except for working ingots to sheet bars, have been limited mainly to forging rocket nozzle and ring shapes. In general, these have been shapes with relatively heavy sections that were produced directly from cast ingots. A typical sequence of operations in forging rings of the Ta-10W alloy is shown in Figure 19-3[6].

For such parts, the ingots are usually heated to 2100-2200 F in gas-fired furnaces adjusted so that the furnace atmosphere is slightly reducing. Because of the large grain size of cast ingots and the oxidation that occurs when the hot billet is exposed to air during forging, the as-forged surface can be quite rough. The typical surface condition that results under these conditions is shown in Figure 19-4. The seams on the heavily-wrinkled surface are generally less than 0.100 inch deep.

To avoid the large material losses associated with poor surface finish, it may be desirable to breakdown the cast ingot structure by first forging between flat dies and recrystallizing the billet. Also, for relatively small closed-die forgings, it may be desirable to use protective coatings to minimize surface contamination.

Two types of coatings have been used with success for protecting tantalum alloys against oxidation during forging — aluminides and glasses. The best protection is provided by aluminum alloy coatings applied by hot dipping. For example, a 3-mil-thick coating of aluminum provided effective protection against oxidation of the Ta-10W alloy heated in air to 2500 F for 30 minutes[7]. A 50Sn-50Al coating on Ta-10W alloy samples has provided good protection against oxidation at temperatures up to 3000 F[8]. In forging at 2150-2400 F, a hot-dipped Al-12Si alloy provided effective protection of the Ta-30Cb-7.5V and Ta-10Hf-5W alloys[4]. This coating is applied by dipping billets in the molten alloy at 1650-1700 F for about 10 minutes.

FIGURE 19-4. SURFACE CONDITION AFTER SAND-BLASTING OF Ta-10W ALLOY PART FORGED DIRECTLY FROM CAST INGOT

Courtesy of Cameron Iron Works

Metallic-coatings, while providing the best oxidation resistance, are generally poor lubricants. Thus, for forging, the use of glass coatings would seem the preferred choice since they offer both protection and lubrication. In the temperature range of 2000 F to 2400 F, a variety of borosilicate glasses are available that would have suitable characteristics (i.e., viscosity, reactivity, etc.).

Forging designs will depend, to some extent, on the condition of the forging billet. Cast billets usually will require more generous allowances for machining to remove surface defects such as seams and shallow laps than wrought-and-recrystallized billets. With wrought stock, designs can be patterned along the lines of stainless steel forgings. However, because die forging experience is limited, materials are expensive, and forging pressures are high, designs should be of more generous contour, particularly in the fillet and corner radii on parts having vertical projections near the edges.

REFERENCES

(1) Schmidt, F. F., and Ogden, H. R., "The Engineering Properties of Tantalum and Tantalum Alloys", DMIC Report 189, Defense Metals Information Center, Battelle Memorial Institute, September 13, 1963.

(2) Schmidt, F. F., "Tantalum and Tantalum Alloys", DMIC Report 133, Defense Metals Information Center, Battelle Memorial Institute, July 25, 1960.

(3) Torti, M. L., "Physical Properties and Fabrication Techniques for the Tantalum 10% Tungsten Alloy", High Temperature Materials II, pp 161-169, Interscience, New York (1963).

(4) Ogden, H. R., et al, "Scale-Up Development of Tantalum-Base Alloys", ASD Technical Report 61-684, Contract AF 33(616)-7452, Battelle Memorial Institute, December, 1961.

(5) Imgram, A. G., and Ogden, H. R., "The Effect of Fabrication History and Microstructure on the Mechanical Properties of Refractory Metals and Alloys", DMIC Report 186, Defense Metals Information Center, Battelle Memorial Institute, July 10, 1963.

(6) "Tantalum Alloy Ingot Reliability and Sheet Rolling Program", First Progress Report - A State-of-the-Art Survey, Contract AF 33(657)-7015, Wah Chang Corporation, November 31, 1961.

(7) Torti, M. L., "Development of Tantalum-Tungsten Alloys for High Performance Propulsion System Components", Contract NOrd 18787, National Research Corporation, July-October, 1960.

(8) Lawthers, D. D., and Sama, L., "Aluminide and Beryllide Protective Coatings for Tantalum", High Temperature Materials II, pp 819-832, Interscience, New York (1963).

CHAPTER 20

MOLYBDENUM AND MOLYBDENUM ALLOYS

Crystalline Structure:	Body-centered cubic
Phase Changes:	None
Number of Phases Present:	One at forging temperatures
Liquidus Temperature:	Unalloyed – 4730 F Mo-W alloys – 4750 to 5400 F Other Mo alloys – 4650 to 4750 F
Solidus Temperature:	Unalloyed – 4730 F Mo-W alloys – 4700 to 5300 F Other Mo alloys – 4700 to 4750 F
Reactions When Heated in Air:	Continuous oxidation above 1000 F; oxide volatile above 1250 F

ALLOYS AND FORMS AVAILABLE

Nominal compositions and forging characteristics of molybdenum and several molybdenum alloys are given in Table 20-1. As a rule, molybdenum alloys may be forged over a wide range of temperatures. Since the alloys have higher recrystallization temperatures than unalloyed molybdenum, the forging temperatures are usually higher.

Molybdenum alloys are prepared commercially by two basic methods: (1) vacuum consumable-electrode arc melting and (2) pressing and sintering. Melting is preferred for preparing the alloys containing reactive elements (e.g., titanium, zirconium, etc.). Billets from arc-melted ingots are generally extruded, and recrystallized to achieve grain refinement before forging. Billets are hot rolled for smaller billet sizes. Pressed and sintered billets are forged directly. Billets of both types of material are available commercially in sizes up to about 6 inches in diameter.

FORGING BEHAVIOR

Molybdenum and molybdenum alloys produced by powder-metallurgy methods normally contain fine grains; arc-melted ingots contain large, radial-columnar grains. These two conditions are illustrated in Figure 20-1.

The influences of billet processing history on forgeability are summarized in Table 20-2. Pressed and sintered billets normally have a density less than theoretical and exhibit some porosity, particularly at the billet centers. The amount of porosity increases with increasing section size. Production experience indicates that forgeability improves with increasing density, and that the minimum density for reasonable forgeability should be at least 90 per cent of theoretical. The average density for

TABLE 20-1. NOMINAL COMPOSITIONS AND FORGING CHARACTERISTICS OF MOLYBDENUM AND SEVERAL MOLYBDENUM ALLOYS

Nominal Composition, wt %	Unalloyed Mo	Mo-0.5Ti	Mo-0.5Ti-0.08Zr	Mo-25W-0.1Zr	Mo-30W
Consolidation Methods[(a)]	AC, PM	AC	AC	AC	AC, PM
Approximate Solidus, F	4730	4700	4700	4800	4900
Approximate Minimum Recrystallization Temp, F	2100	2400	2600	2600	2300
Minimum Hot-Working Temp[(b)], F	2400	2700	3000	3000	2500
Normal Forging Temp Range, F	1900-2400	2100-2600	2200-2700	1900-2400	2100-2400
Lowest Reported Forging Temp, F	1100	1800	1200	1400	2200
Forgeability					
Presses	Good	Fair	Good	Fair	--
Hammers	Good	Good	Good	--	Fair
High-energy-rate machines	Good	--	Good	--	--

(a) AC - consumable-electrode arc-cast; PM powder-metallurgy methods.

(b) Minimum hot-working temperature is defined as the lowest temperature at which the alloy will begin to recrystallize during forging. Below these temperatures, the alloys are cold worked.

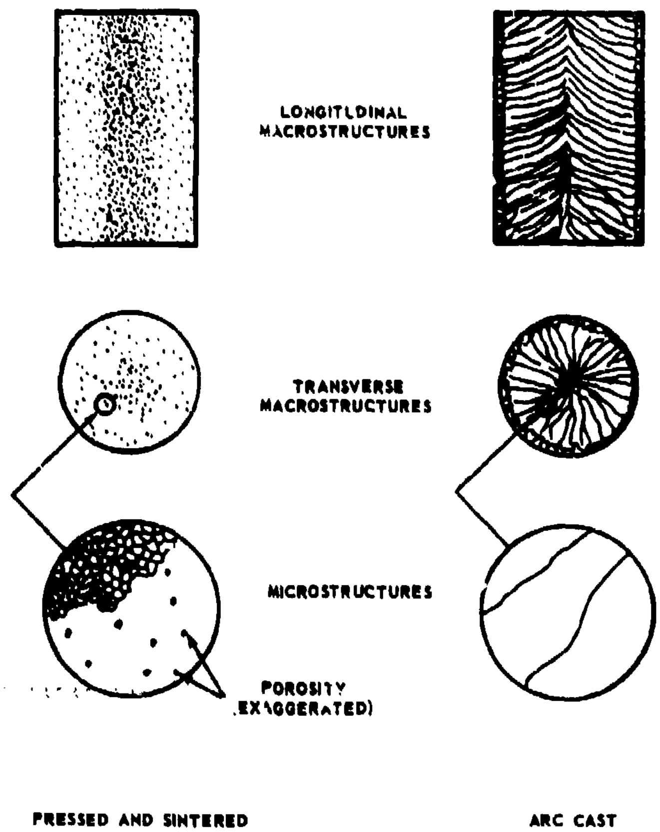


FIGURE 20-1. TYPICAL MACROSTRUCTURES AND MICROSTRUCTURES OF PRESSED-AND-SINTERED AND ARC-CAST MOLYBDENUM

billets larger than 4 inches in diameter should be at least 92 per cent of theoretical, since the density decreases from surface to center.

TABLE 20-2. INFLUENCE OF BILLET HISTORY ON FORGEABILITY OF MOLYBDENUM AND MOLYBDENUM ALLOYS[1]

Billet Condition	Comments on Forgeability
Pressed and sintered	Generally forgeable in any direction; billet centers become weaker with increasing diameter because of decreasing density; large billets are either extruded or side forged before die forging
Arc cast	Not forgeable except at very high temperatures (3500 F or higher)
Arc cast and extruded	Forgeability improves with increasing extrusion ratio; ratio of 4:1 is about minimum for reasonable forgeability
Arc cast, extruded, and recrystallized	Forgeable in any direction; billets extruded with ratios of less than 2:1 do not always recrystallize completely during subsequent annealing; hence, they often show reduced upset forgeability

Because of their coarse, radially oriented grains, arc-melted ingots are generally brittle when subjected to tensile deformation. For this reason, ingots are normally extruded at warm-working temperatures with at least a 2:1 ratio. The most common practice for achieving good forgeability consists of extrusion at hot-cold-working temperatures with at least a 4:1 extrusion ratio, and recrystallization of the extruded bar to refine the grain size. The forgeability of arc-cast ingots improves at true hot-working temperatures, but such procedures have not proven as reliable in practice as extrusion breakdown. Furthermore, in closed-die forging studies on the Mo-0.5Ti-0.08Zr alloy, Westinghouse found that billet stock produced by direct forging at true hot-working temperatures did not have as good forgeability as extruded and recrystallized billet stock.[2]

Wrought molybdenum and molybdenum-alloy billets with a fine grain structure may be forged successfully with large reductions over a wide range of temperatures and deformation rates. This has been demonstrated in forging studies on arc-cast-and-extruded billets of unalloyed molybdenum at Lockheed[3] and the Mo-0.5Ti-0.08Zr alloy at Battelle[4]. A summary of the conditions under which these materials could be successfully forged is given below:

Alloy	Temperature Range, F	Forging Rates, in./sec	Maximum Reduction Without Cracking, %
Unalloyed Mo	1100-2200	1.0 to 400	89
Mo-0.5Ti-0.08Zr	1200-2000	0.1 to 200	94
Mo-0.5Ti-0.08Zr	2500-3000	0.1 to 200	50

Forgeability of both materials was virtually unaffected by drastic changes in the forging rate. Gross changes in forging temperature in the hot-cold-working region similarly had no noticeable effect on forgeability. However, forgeability is markedly reduced at high temperatures in the hot-cold-working region approaching true hot-working conditions, as indicated by the data for the Mo-0.5Ti-0.08Zr alloy. The Mo-25W-0.1Zr alloy exhibits the same type of forging behavior.[5,6] Reductions as high as 90 per cent can be achieved at temperatures in the range 1400 F to 1900 F, but forgeability becomes progressively poorer at higher temperatures, particularly above about 2400 F. The lowest practical forging temperature for these alloys (for such operations as finishing, sizing, coining, etc.) appears to be about 1000 F.

Table 20-3 lists temperatures for forging, recrystallization, and annealing several commercial and experimental molybdenum-base alloys. The forging temperatures given represent normal ranges reported by industry; however, these temperatures are sometimes adjusted to obtain special properties. For example, higher forging temperatures are used to maintain higher recrystallization temperatures; lower forging temperatures are used to develop higher room-temperature strengths. In general, most alloys are work hardened during forging, since the forging temperatures listed in Table 20-3 are lower than the minimum hot-working temperatures.

Table 20-4 gives average forging pressure and forgeability data for upsetting unalloyed molybdenum and several molybdenum alloys at a constant temperature. The rapid increases in pressure with increasing reduction demonstrate the work-hardening characteristics.

The pressure and specific energy (work per unit volume) required for forging molybdenum alloys are not influenced greatly by differences in either temperature or forging rate in the hot-cold-working range. Upset forging pressures for the Mo-0.5Ti-0.08Zr alloy are presented in Figure 20-2 for reductions up to 30 per cent at temperatures between 1500 and 3000 F at two strain rates characteristic of normal press-forging operations. The forging pressures at temperatures between 1500 and 2500 F are almost the same. At 3000 F, however, the pressures are slightly lower, particularly at the lower strain rate. The influence of forging temperature on forging pressure is illustrated in Figure 20-3. Hot-working conditions are approached only when the Mo-0.5Ti-0.08Zr alloy is forged at 3000 F. The forging-pressure behavior for two steels is shown in Figure 20-3 for comparison.

The essentially constant specific-energy requirements for upsetting unalloyed molybdenum and Mo-0.5Ti-0.08Zr alloy over a wide range of temperatures are shown in Figure 20-4. By comparison, data obtained similarly for A-286 and AISI 4340 steels show a marked reduction in the specific forging energy with increasing temperatures. It is interesting to note that, at 1800 F, the specific energy required for 50 per cent upset is essentially the same for both the A-286 and the Mo-0.5Ti-0.08Zr alloys.

When forging molybdenum, it is important to realize that reheating a blank during forging in the hot-cold-working range does not remove the effects of cold work imparted by preceding forging reductions. The data given in Table 20-5 show that pressures required for reforging previously forged blanks are nearly equal to the forging pressures for the previous reduction. These data show again that forging pressures are influenced only slightly by large temperature changes in the warm-working region.

TABLE 20-3. TYPICAL FORGING, RECRYSTALLIZATION, AND STRESS-RELIEF TEMPERATURES FOR SEVERAL COMMERCIAL AND EXPERIMENTAL MOLYBDENUM ALLOYS[1]

Data Based on Industry Survey

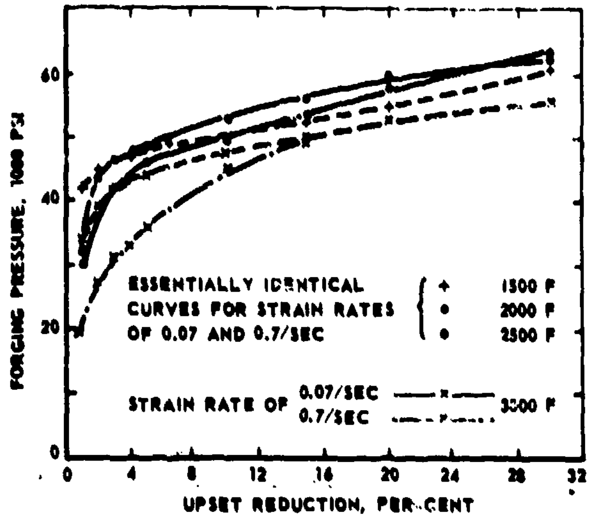
Alloy	Forging Temperatures, F: Initial Breakdown	Forging Temperatures, F: Final Forge	Recrystallization Temperatures of Forgings, F	Stress-Relief Annealing Temperatures, F
		Commercial Alloys		
Unalloyed molybdenum				
Pressed and sintered	2100-2400	2000-2200	2100-2200	1700-2000
Arc cast and extruded	2100-2400	1900-2200	2000-2200	1700-2000
Mo-0.5Ti	2300-2600	2100-2400	2300-2500	2000-2200
Mo-0.5Ti-0.08Zr	2400-2700	2200-2500	2700-3000	2200-2400
Mo-30W	2200-2400	2100-2300	2300-2500	1900-2000
		Experimental Alloys		
Mo-25W-0.1Zr	2000-2400	1900-2300	2800-3100	2200-2400
Mo-0.25Zr	2400-2700	2400	3000-3200	--
Mo-0.5Zr	2400	--	2800-3000	--
Mo-1.5Cb	2400	--	2400-2800	--
Mo-1.2Ti-0.25Zr-0.15C	2700	--	over 3100	--

TABLE 20-4. AVERAGE FORGING PRESSURE FOR EXTRUDED AND RECRYSTALLIZED MOLYBDENUM-ALLOY BARS FORGED AT 2400 F[5]

Single Upset in a 700-Ton Press at 80 ipm

Alloy	Initial Deformation Pressure, psi	Final Pressure, psi, to Achieve Reduction in Height Indicated: 45-55 Per Cent	56-65 Per Cent	66-75 Per Cent	Greatest Reduction Achieved Without Fracture, per cent
Unalloyed molybdenum	22,000	64,000	91,000	--	70
Mo-0.5Ti-0.08Zr	39,000	95,000	118,000	132,000	83
Mo-25W-0.1Zr	60,000	126,000	186,000	--	55
Mo-1.5Cb	30,000	99,000	109,000	--	60
Mo-0.5Zr	32,000	88,000	95,000	--	69
Mo-0.25Zr	27,000	83,000	97,000	--	66

FIGURE 20-2.
FORGING-PRESSURE CURVES FOR UPSETTING Mo-0.5Ti-0.08Zr ALLOY AT VARIOUS TEMPERATURES AND STRAIN RATES[4]



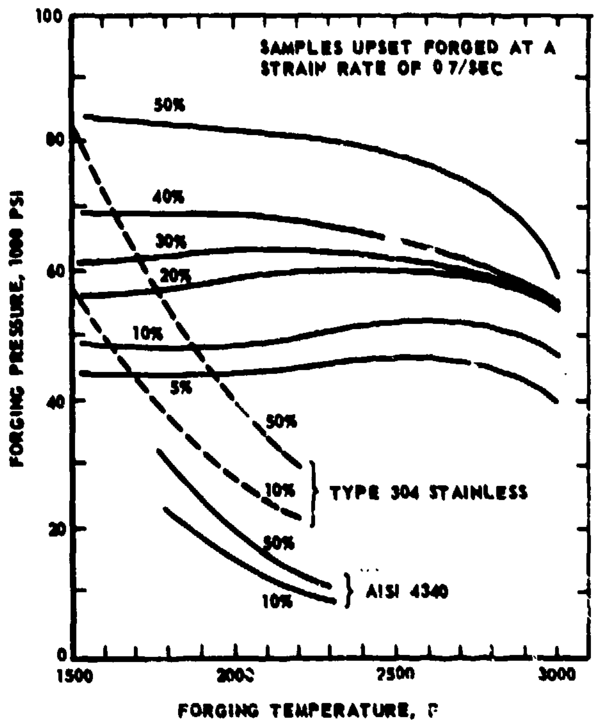


FIGURE 20-3.
EFFECT OF TEMPERATURE ON FORGING PRESSURE FOR Mo-0.5Ti-0.08Zr ALLOY UPSET TO VARIOUS REDUCTIONS[4]

SIMILAR DATA FOR AISI 4340 AND TYPE 304 STAINLESS ARE GIVEN FOR COMPARISON

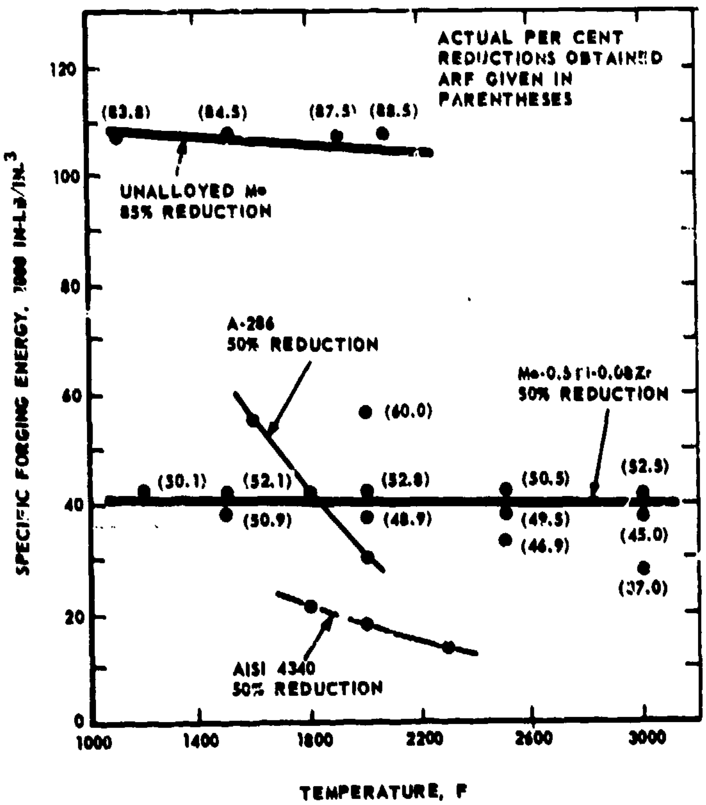


FIGURE 20-4. INFLUENCE OF TEMPERATURE ON THE SPECIFIC FORGING ENERGY FOR UPSETTING MOLYBDENUM AND Mo-0.5Ti-0.08Zr ALLOY TO ESSENTIALLY CONSTANT REDUCTIONS IN A DYNAPAK[3,4]

DATA FOR AISI 4340 AND A-286 STEELS SHOWN FOR COMPARISON

TABLE 20-5. INFLUENCE OF PRIOR COLD WORK ON THE FORGING PRESSURE OF Mo-0.5Ti-0.08Zr ALLOY[4]

After initial 50 per cent upset, samples remachined to cylinder with original diameter of 1.5 inches. Upset at 80 ipm.

Temperature, F	Average Forging Pressure, 1000 psi, at Given Per Cent Upset Reductions						
	First Upset[a]				Second Upset		
	5%	10%	20%	50%	5%	10%	20%
2000	44	50	57	81	76	84	96
2500	48	53	57	[illegible]	74	79	90

(a) Samples from arc-cast, extruded, and recrystallized bar.

INFLUENCE OF FORGING VARIABLES ON MECHANICAL PROPERTIES

The strength of molybdenum-alloy forgings is developed primarily by work hardening. Thus, the control of mechanical properties requires carefully planned forging sequences. These can be readily established since molybdenum exhibits classical strain-hardening behavior, i.e., hardness increases with increasing reduction, and decreases with increasing forging temperature.

There are limits, however, to which these alloys should be cold worked to achieve a desired level of elevated-temperature properties. The amount of cold work influences the recrystallization behavior of the alloys which, in turn, establishes the maximum service temperatures. Hence, the effect of forging variables on recrystallization behavior must be taken into account in planning a forging sequence. At a given forging temperature, the recrystallization temperature generally decreases with increasing reduction. Conversely, at a given level of reduction, the recrystallization temperature generally increases with increasing forging temperature. This is illustrated by the data given in Table 20-6, which shows the influence of forging conditions on the recrystallization behavior of the Mo-0.5Ti-0.08Zr alloy. These data show that the recrystallization temperatures of the alloy may vary as much as 600 F, depending on forging conditions. The lowest annealing temperature for complete recrystallization of the Mo-0.5Ti-0.08Zr alloy is about 2500 F. This was observed for sheet rolled to reductions exceeding 90 per cent at temperatures below 2200 F.[7]

<u>Hardness</u>. A method for studying the work-hardening behavior of molybdenum alloys is the wedge-forging test illustrated in Figure 20-5. This test consists of forging a wedge-shaped blank to a constant thickness at various temperatures. After being forged, the sample contains material with constantly increasing linear reductions varying from zero to about 65 per cent along the sample length. Thus, this test provides quantitative data on the influences of forging temperature and forging reduction on hardness.

TABLE 20-6. INFLUENCE OF FORGING TEMPERATURE AND FORGING REDUCTION ON THE RECRYSTALLIZATION BEHAVIOR OF THE Mo-0.5Ti-0.08Zr ALLOY[4,8]

Forging Temperature, F	Per Cent Reduction	Approximate 1-Hour Annealing Temperature for Recrystallization, F: Recrystallization Begins	Recrystallization Complete
2200	30	2600	2800
	60	2500	2700
2400	10	--	3000
	30	2600	2800
	60	--	2700
2900	10	2800	3200
	30	2600	3000
	60	--	2700
3400	10	3000	3200
	30	2800	3000
	60	2700	2800

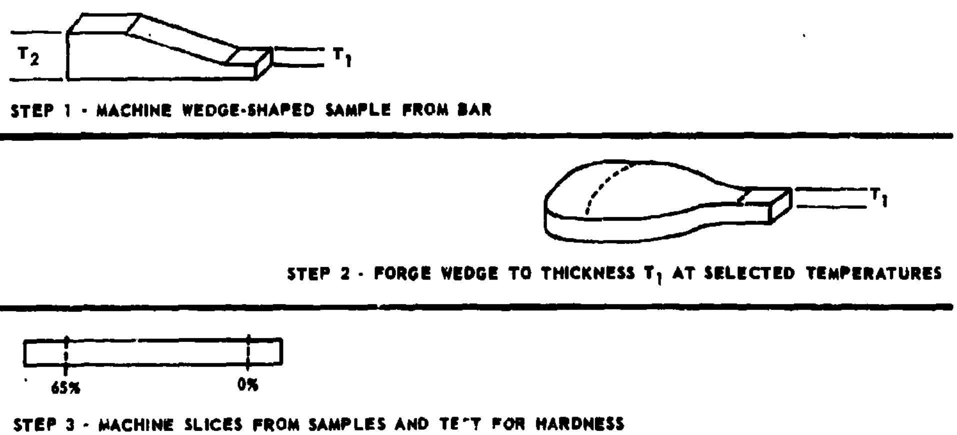


FIGURE 20-5. SEQUENCES FOR THE WEDGE-FORGING TEST

Hardness curves for unalloyed molybdenum and several molybdenum alloys determined by wedge-forging tests at various temperatures are shown in Figure 20-6. Unalloyed molybdenum and the Mo-0.5Zr alloy exhibit classical work-hardening behavior. Alloys containing titanium, however, exhibit a form of strengthening in addition to strain hardening. This is evident from the hardness increases with increasing forging temperature. It is believed that the hardening is a result of an accompanying aging reaction. This is suggested by studies at General Electric[9] which indicate an age-hardening response attributed to an "extensive precipitation of titanium carbide" at temperatures in the vicinity of 2800 F. Other investigations[4,7,8,10] have described a similar hardening response for titanium-bearing alloys between 2600 F and 3000 F. This behavior appears to be the reason for the slight increase in forging pressure noted earlier for the Mo-0.5Ti-0.08Zr alloy when forging temperatures were increased from 2000 F to 2500 F (Figure 20-3).

Over the range of forging temperatures from 2700 F to 3450 F, the Mo-25W-0.1Zr alloy exhibits progressively smaller hardness increases with increasing temperatures. This alloy also appears to exhibit an aging response, since the hardnesses at low reductions are considerably higher than those for the starting blanks.

<u>Tensile Properties</u>. The following points summarize the influence of forging procedure on the tensile properties of molybdenum forgings:

(1) Strength and hardness decrease with increasing forging temperature, but are not significantly affected by the forging rate. Table 20-7 gives the mechanical properties of unalloyed molybdenum upset forged between 1100 and 2200 F at high forging rates in a Dynapak. Forging to comparable reductions in both a drop hammer and a hydraulic press at the same temperatures produced the same hardness levels and, thus, comparable strength levels.

TABLE 20-7. INFLUENCE OF FORGING TEMPERATURE ON HARDNESS AND STRENGTH OF UNALLOYED MOLYBDENUM[3]

Upset forged in a Dynapak from arc-cast, hot rolled, and recrystallized bar.

				Room-Temp Mechanical Properties		
Forging Temp, F	Forging Reduction, per cent	Avg Hardness R_A	Avg Hardness DPH	Yield Strength, 1000 psi	Ultimate Strength, 1000 psi	Elongation, per cent
As Received	--	56-57	--	76	86	24.9
1100	84	62	263	112	123	9
1500	84	62	252	109	124	14
1900	87	61.5	246	102	112	18
2200	90	59	226	[illegible]	103	24

(2) Parts forged with large reductions usually exhibit anisotropic properties. Highest strengths are obtained in directions parallel to the direction of greatest metal flow, as indicated by these typical room-temperature tensile properties for molybdenum rolled-ring forgings:

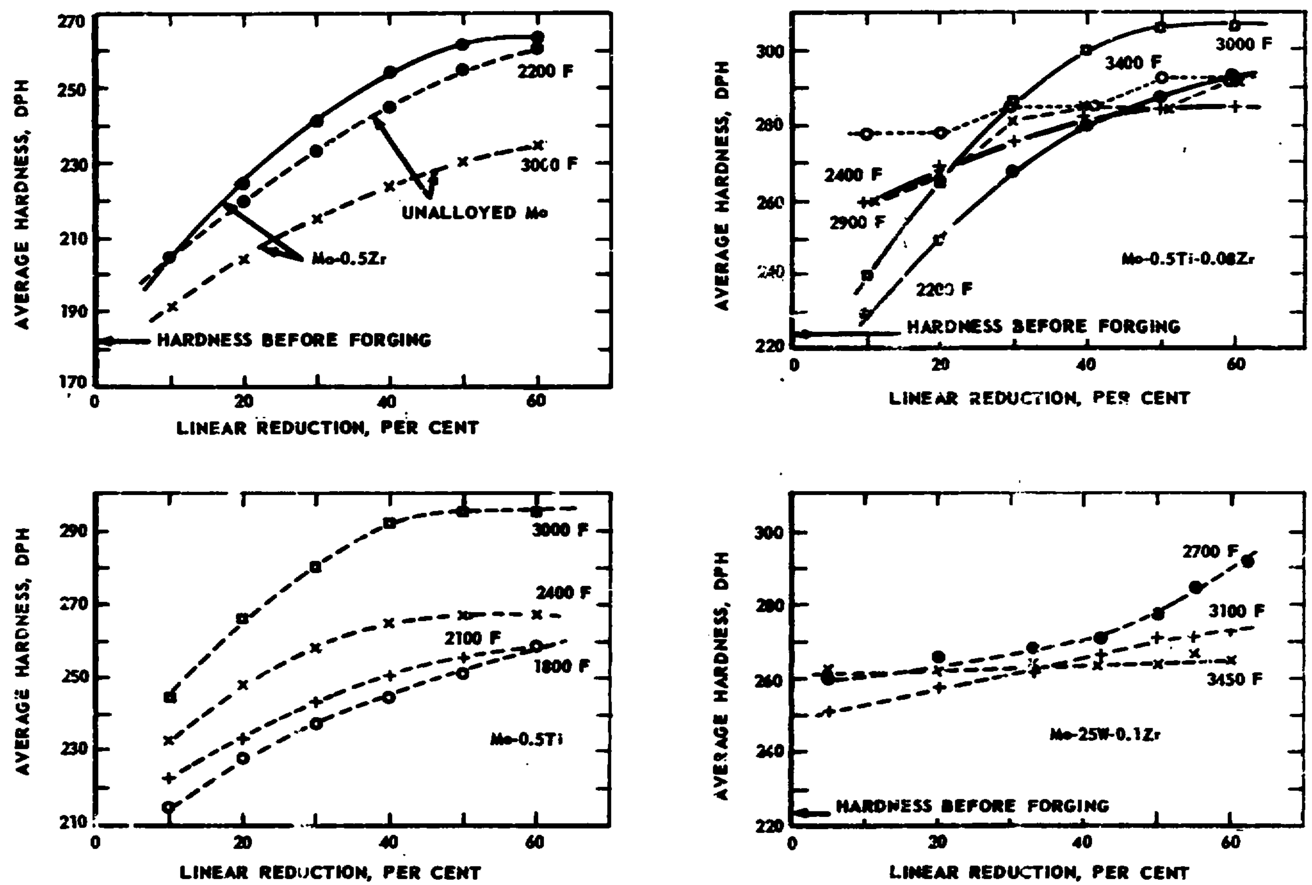


FIGURE 20-6. EFFECT OF REDUCTION ON HARDNESS OF MOLYBDENUM AND MOLYBDENUM ALLOYS DETERMINED BY WEDGE-FORGING TESTS AT VARIOUS TEMPERATURES(5,8)

Direction	Yield Strength, 1000 psi	Ultimate Strength, 1000 psi	Elongation, per cent	Reduction of Area, per cent
Transverse	73-75	83-84	38-42	56-62
Longitudinal	87-91	90-97	16-22	20-42

(Data courtesy of Aerojet General Corporation.)

(3) Forgings containing regions given varying amounts of reduction exhibit strength properties that vary accordingly. These room-temperature strength properties are typical of large molybdenum nozzle forgings which contain both a severely deformed flared section and a lightly worked apex section:

Location	Yield Strength, 1000 psi	Ultimate Strength, 1000 psi	Elongation, per cent	Reduction of Area, per cent
Flared section	75-81	86-92	11-29	16-43
Apex	65-69	71-81	7-19	11 20

Mechanical properties given represent three heats of arc-cast and extruded unalloyed molybdenum forged at 2200 F and stress relieved for 1/2 hr at 1900 F.
(Data courtesy of Cameron Iron Works.)

(4) Parts forged with severe reductions will recrystallize at lower temperatures than parts given light reductions. For example, thin molybdenum blade forgings may recrystallize when heated to temperatures as low as 2000 F, while large cones will not recrystallize until heated to temperatures over 2200 F. Thus, careful control of forging temperatures and reductions is necessary to avoid premature recrystallization in service with an attendant loss of strength.

Impact Toughness. Close control of the forging operation to avoid recrystallization is important in achieving good impact properties in unalloyed molybdenum parts. Residual carbon in arc-cast molybdenum forms grain-boundary carbides which affect ductility adversely. Roshong and Leeper[11] found that unnotched impact properties equivalent to those attainable with powder-metallurgy molybdenum billets could be obtained with arc-cast molybdenum containing less than 50 ppm of carbon and correspondingly low oxygen. With both types of materials, however, it was necessary to use the following forging practice:

(1) Forge with at least 20 per cent reduction in the finish forging operation

(2) Finish forge at a temperature sufficiently low to avoid self-recrystallization

(3) Stress-relief anneal below the recrystallization temperature.

The effect of recrystallization on the impact and tensile properties of unalloyed molybdenum forgings is shown in Figure 20-7[11]. It can be seen that impact toughness is reduced quite markedly at the very onset of recrystallization. On the other hand, strength and ductility do not significantly change until gross recrystallization has occurred.

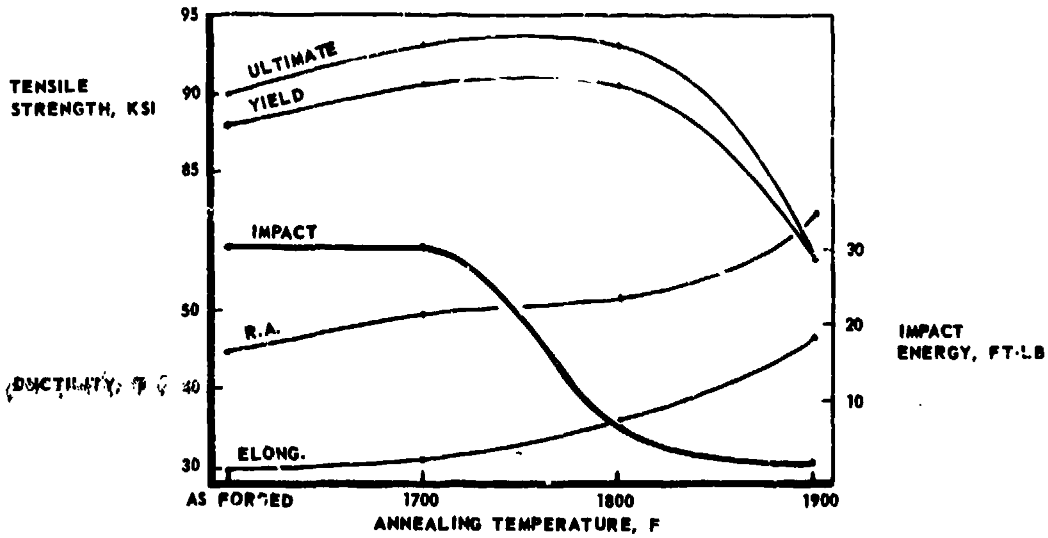


FIGURE 20-7. EFFECT OF ANNEALING TEMPERATURE ON IMPACT AND TENSILE PROPERTIES OF UNALLOYED MOLYBDENUM FORGINGS[11]

Control of Properties. The control of properties in die forgings can be illustrated by planning the forging sequences for a hypothetical part:

Example. A molybdenum die forging is designed in such a way that it will require forging reductions varying between about 30 per cent and 80 per cent. The forging contains thin sections that are characterized by predominant longitudinal grain flow. The part requires one open die and two closed-die forging operations (draw one end, block, and finish) and three heating operations.

Problem. How to obtain uniform mechanical properties and recrystallization behaviors throughout the forging?

Solution. Referring to the hardness curves in Figure 20-6, it is apparent that hardness increases with increasing reduction at any given forging temperature. Also, increasing reductions will reduce the recrystallization temperature of molybdenum. These two effects are illustrated by the schematic diagram shown in Figure 20-8. Since the hypothetical

forging will receive between 30 and 80 per cent reduction, its hardness and recrystallization temperature will vary widely. To avoid this variation, it is desirable to use an in-process recrystallization treatment, probably after the drawing-out operation. This would remove the effects of differential reductions during drawing. During subsequent die forging, the part would then be given effective reductions varying only between about 40 and 60 per cent. In this way, the part would exhibit more uniform recrystallization and mechanical-property behavior than if forged directly from the starting billet. As an added benefit, the in-process annealing treatment would reduce property directionality in regions that otherwise would have strongly oriented grain flow.

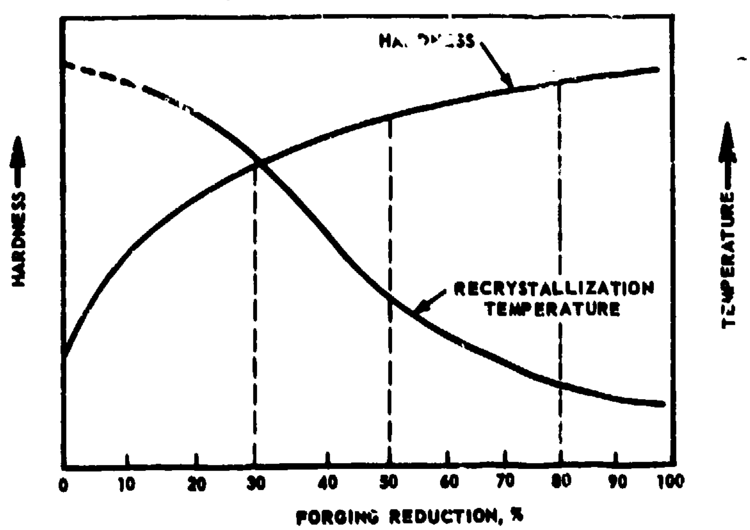


FIGURE 20-8. RELATIVE INFLUENCE OF FORGING REDUCTION ON HARDNESS AND RECRYSTALLIZATION BEHAVIOR OF COLD-WORKED MOLYBDENUM

Essentially this procedure is used in forging practice for such parts. However, molybdenum recrystallizes at temperatures obtainable in conventional furnaces. For this reason, the most common practice consists of performing initial forging operations at about 2200 F, where recrystallization occurs during forging. Final forging operations are then conducted at lower temperatures (1900 to 2100 F) to provide the necessary work hardening. These procedures would be considered for alloys having recrystallization temperatures above about 2300 F.

COMMENTARY ON FORGING PRACTICES

The metallurgical principles involved in the forging of molybdenum are similar to those for the work-hardenable austenitic stainless steels and superalloys (e.g., 16-25-6, 19-9-DL). Principles common to both systems are:

(1) Cold work is necessary to increase strength.
(2) Recrystallization temperatures decrease with increasing cold work.
(3) Forgeability increases with decreasing grain size.
(4) Forging pressure increases with increasing reduction.
(5) Forgeability is not influenced greatly by deformation rate.
(6) Die-chilling effects place practical limits on section sizes obtainable in die forgings.

About the only significant difference between the two systems is their oxidation behavior which influences heating, lubricating, and handling procedures.

Heating. At forging temperatures above about 1400 F, molybdenum forms a volatile oxide so fast that surface contamination is rarely a problem. For this reason, conventional oil- and gas-fired furnaces may be used for heating to temperatures in the vicinity of 2500 F. Argon, hydrogen, or carbon monoxide atmospheres are used when metal losses become excessive. Higher forging temperatures require electric-resistance or induction furnaces. In these cases, the use of protective atmospheres is increasingly important to protect the billet from oxidation and the furnace from attack by the oxide vapors. Molybdenum oxide vapor (1) attacks thermocouples, causing erroneous temperature readings, (2) reacts with electric-resistance heating elements, causing brittleness and premature failure, and (3) attacks refractory brick, lowering its refractoriness. For these reasons it is helpful to "burn-off" a furnace for 1 or 2 hours after the last piece of molybdenum is removed. This may be done by holding the furnace at about 2200 F and allowing air circulation to flush out the vapor.

Lubricants. The liquid oxide formed on molybdenum during heating serves as an excellent lubricant. However, to reduce metal loss, glass-type billet coatings are often used on large forgings. Glass coatings also reduce heat losses during forging. Colloidal graphite and molybdenum disulfide are suitable die lubricants for small forgings.

Die Materials. For hammer-forging dies, low-alloy tool steels (e.g., Ladish D6, Heppenstall Hardtem, Finkel FX) and medium-alloy Type H-11 tool steels are most commonly used. These same tool steels are used for press-forging dies. For longer die life, inserts of Types H-12 and H-13 tool steels are sometimes recommended for press-forging dies.

Cooling from Forging and Stress-Relief Annealing. Forgings are usually either air cooled from forging temperatures or charged directly into a stress-relief furnace. Some of the alloys are cooled in an insulating material to prevent stress cracking. When forgings are cooled very slowly through the range of 1500 F to 1100 F, however, the liquid oxide can penetrate the grain boundaries and cause a "network". Although this "network" is generally shallow, it presents an unattractive surface appearance that can complicate the subsequent inspection for major defects.

Forgings are usually given a stress-relief annealing treatment at a temperature about 100 F below the recrystallization temperature. This treatment improves both room-temperature ductility and machinability. Specific 1-hour annealing temperatures reported for forgings are:

Unalloyed Mo	1600 - 1800 F
Mo-0.5Ti	2000 F
Mo-0.5Ti-0.08Zr	2000 - 2200 F
Mo 25W-0.1Zr	2000 - 2150 F
Mo-30W	2100 - 2300 F

Since the ductile-brittle transition temperatures of most molybdenum alloys are close to room temperature, cool forgings should be handled carefully.

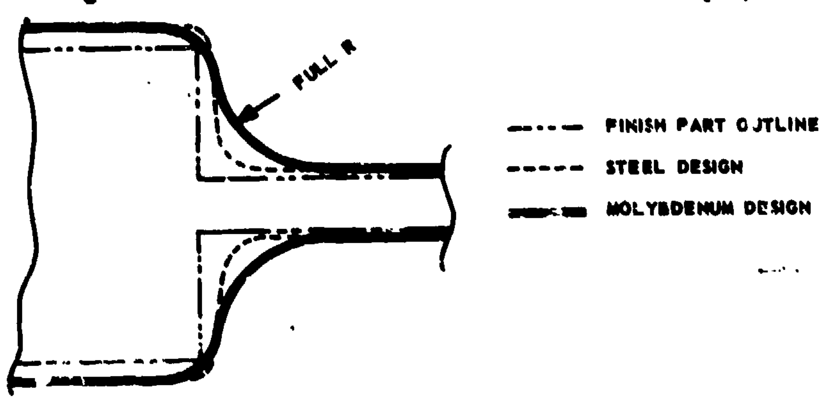
COMMENTARY ON FORGING DESIGN

Since the metallurgical quality of molybdenum and molybdenum alloy forgings depend greatly on the control of forging reductions, the designs are more critical than those for aluminum or steels forgings. The designs should be more generously contoured to provide both smooth metal flow and more readily controlled reductions.*

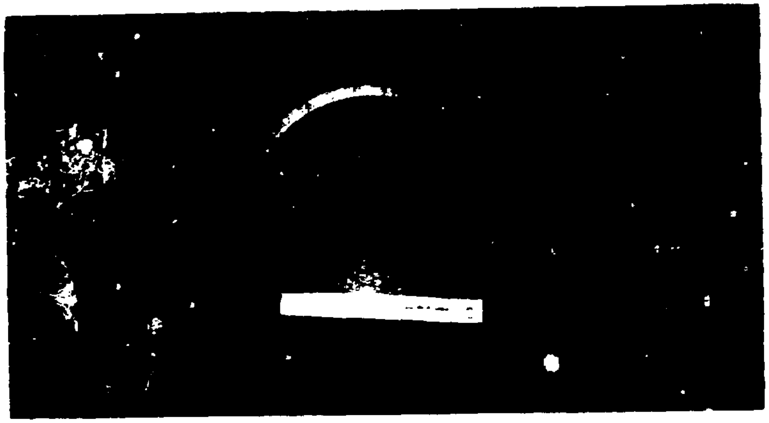
The size of molybdenum forgings is limited somewhat by the billet sizes available. One of the largest forgings produced is a 150-pound nozzle cone about 12 inches in diameter by 12 inches long. This was forged from an 8-inch-diameter pressed-and-sintered billet.

Forging Shapes. The greatest die-forging experience has been with circular shapes (e.g., cones, rings, disks) varying in size up to about 16-inch diameter, but structural shapes, blades, and shafts also have been successfully forged. Some typical examples of molybdenum die forging are shown in Figures 20-9 and 20-10.

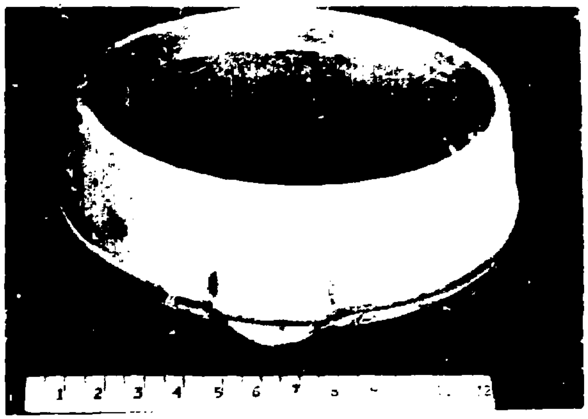
Design Details. Since molybdenum alloys require greater forging pressures than low-alloy and stainless steels, it is good practice to use more generous fillet and corner radii, and allow for more die wear than for steel. The draft angles necessary for molybdenum forgings are about the same as those for austenitic stainless steels (5 to 10 degrees). When desired shapes contain abrupt changes in section sizes, it is good practice to use generous radii between such areas. As an example, the sketch below



*Forging-design practices for molybdenum and its alloys are also described in Chapter 4.



Courtesy of Canton Drop Forging and Manufacturing



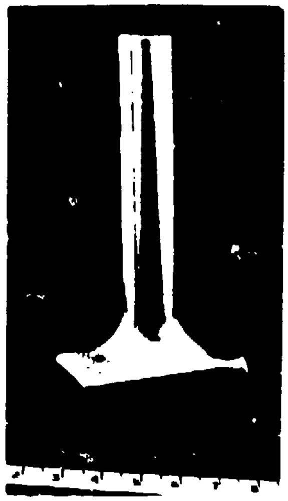
Courtesy of Wyman-Gordon C.

Courtesy of Wyman-Gordon Co.

Courtesy of Steel Improvement and Forge Company

FIGURE 20-9. TYPICAL UNALLOYED MOLYBDENUM DIE FORGINGS PRODUCED FROM BOTH ARC-CAST-AND-EXTRUDED AND PRESSED-AND-SINTERED BILLETS

shows the preferred method of designing a molybdenum part containing a 2-1/2-inch section adjacent to a 1/2-inch section by comparing it with the normal design for steel. This approach allows for a more gradual change in the amount of reduction during forging. Greater reproducibility of both dimensions and properties can be expected because the more-generously contoured designs require fewer forging steps.



Courtesy of Wyman-Gordon Company

Courtesy of Cameron Iron Works

FIGURE 20-10. TYPICAL UNALLOYED MOLYBDENUM DIE FORGINGS PRODUCED FROM ARC-CAST-AND-EXTRUDED BILLET

Section Thickness. The combination of high forging-pressure requirements and rapid cooling rates imposes limitations on the section size obtainable in molybdenum die forgings in various types of forging equipment. Practical minimum section-size limitations for molybdenum and low-alloy steel forgings are compared in Table 20-8. It

TABLE 20-8. COMPARISON OF MINIMUM PRACTICAL SECTION SIZES FOR FORGED MOLYBDENUM ALLOYS AND LOW-ALLOY STEELS

Type of Forging	Hydraulic Presses		Drop Hammers		High-Energy-Rate Equipment[a]	
	Mo	Steel	Mo	Steel	Mo	Steel
Blades	1/4	1/4	1/8	1/8	3/32	3/32
Disks						
Less than 6-in. diam	1/2	1/2	1/4	1/4	1/8	3/32
6 to 12-in. diam	5/8	1/2	3/8	3/8	--	--
Over 12-in. diam	3/4	5/8	1/2	1/2	--	--
Spars[b]						
Less than 3-in. long	1/2	1/2	3/8	1/4	--	--
6 to 12-in. long	1/2	1/2	1/2	3/8	--	--
Over 12-in. long	5/8	5/8	1/2	1/2	--	--

(a) The values listed have been approximated from data obtained from the Bendix Corporation and from the General Dynamics Corporation.

(b) For this comparison, [illegible]s are defined as I-sections tapered end-to-end and having a width of about 1/3 the length.

should be recognized that these minimum section sizes are producible in forgings of rather simple shape. For increasingly complex configurations, these values should be looked upon only as target minimum thicknesses.

Tolerances. As a general rule, close tolerances on die forgings require added operations such as coining or restriking. These added operations usually result in a greater scatter of mechanical properties in molybdenum forgings as well as added processing costs. While it is possible to achieve dimensional tolerances in the vicinity of ±0.020 inch on typical forgings, tolerances of ±0.060 inch would insure greater reproducibility and minimize the number of forging operations. The closer tolerances might be recommended for parts where there is a potential for large quantities of forgings.

Tolerances suggested for molybdenum-alloy disks of varying diameters are:

Disk	Tolerances, inch	
Diameter, inches	Die Closure	Diameter
3	±0.040	±0.010
5	±0.060	±0.020
10	±0.080	±0.030
20	±0.120	±0.060

The suggested die-closure tolerances are about double those recommended for low-carbon steels; the diametral tolerances are about the same.

REFERENCES

(1) Henning, H. J., and Boulger, F. W., "Notes on the Forging of Refractory Metals", DMIC Memorandum 143, Defense Metals Information Center, Battelle Memorial Institute (December 21, 1961).

(2) Goldenstein, A. W., "Molybdenum Forging Process Development", Technical Documentary Report ML-TDR-64-41, Contract AF 33(600)-41419, Westinghouse Electric Corporation (February, 1964).

(3) Pipher, F. C., "Molybdenum High Energy Rate Precision Forging Development", Lockheed Aircraft Corporation Report LR 15290 (July 31, 1961).

(4) Henning, H. J., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", Technical Documentary Report ML-TDR-64-95, Contract AF 33(600) 42963, Battelle Memorial Institute (March, 1964).

(5) Tombaugh, R. W., et al., "Development of Optimum Methods for the Primary Working of Refractory Metals", WADD Technical Report 60-418, Contract AF 33(616)-6377, Harvey Aluminum Inc. (August, 1961).

(6) Carnahan, D. R., and Visconti, J. A., "The Extrusion, Forging, Rolling, and Evaluation of Refractory Alloys", Technical Documentary Report ASD-TDR-62-670, Contract AF 33(616)-8325, Westinghouse Electric Corporation (October, 1962).

(7) Semchyshen, M., and Barr, R. Q., "Investigation of the Effects of Hot-Cold Work on the Properties of Molybdenum Alloys", WADC Technical Report 56-454, Contract AF 33(616)-2861, Climax Molybdenum Company (January, 1957).

(8) Semchyshen, M., and Barr, R. Q., "Effects of Thermal Mechanical Variables on The Properties of Molybdenum Alloys", WADD Technical Report 60-451, Contract AF 33(616)-5441, Climax Molybdenum Company (June, 1960).

(9) Chang, W. H., "A Study of the Influence of Heat Treatment on Microstructure and Properties of Refractory Alloys", Technical Documentary Report, ASD-TDR-62-211, Contract AF 33(616)-7125, General Electric Company (April, 1962).

(10) Semchyshen, M., McArdle, G. D., and Barr, R. Q., "Development of High Strength and High Recrystallization Temperatures on Molybdenum-Base Alloys", WADC Technical Report 58-551, Contract AF 33(616)-2861, Climax Molybdenum Company (September, 1958).

(11) Roshong, R. L., and Leeper, W. A., "Production of Unalloyed Molybdenum Forgings For Service Requiring High Impact Strength", paper presented at the Symposium on High-Temperature Refractory Metals, AIME Annual Meeting, Los Angeles, California (December 9-10, 1963).

CHAPTER 21

TUNGSTEN AND TUNGSTEN ALLOYS

Crystalline Structure:	Body-centered cubic
Phase Changes:	None
Number of Phases:	One at forging temperatures
Liquidus Temperature Range:	Near 6170 F
Solidus Temperature Range:	5000-6100 F
Reaction When Heated in Air:	Continuous oxidation above 1000 F; oxide volatile above 1400 F

ALLOYS AND FORMS AVAILABLE

Table 21-1 gives the compositions and general characteristics of tungsten and several tungsten alloys that have been forged.(1-4) Tungsten-base materials, like the other refractory alloy systems, can be classified into two broad groups:

(1). Unalloyed metal and solid-solution alloys (typified by additions of molybdenum or rhenium)

(2) Dispersion-strengthened alloys (typified by additions of thoria).

This classification is convenient because it categorizes the alloys not only according to their metallurgical behavior, but also by the applicable consolidation methods. Thus, while the solid-solution alloys and unalloyed tungsten can be produced by either powder-metallurgy or melting techniques, forging billets of ThO_2 dispersion-strengthened alloys can be produced only by powder-metallurgy methods.

TABLE 21-1. GENERAL CHARACTERISTICS OF TUNGSTEN AND SEVERAL TUNGSTEN ALLOYS(1,2,3,4)

Nominal Composition, %	Consolidation Methods(a)	Approximate Solidus, F	Typical Forging-Temperature Range, F	Typical Recrystallization Temperature, F	Typical Stress-Relief Temperature, F
Unalloyed W	PS, AC, EB, PAS	6170	2200-3000	2600-2900	2200-2400
W-1ThO_2	PS	6170	2400-3500	2900-3000	2[illegible]0-3400
W-2ThO_2	PS	6170	2400-[illegible]	3000-3200	[illegible]
W-2Mo	AC, PS	6125	2200-2500	2800-3000	[illegible]
W-15Mo	AC, PS	5970	2000-2500	2700-2900	2300-2500
W-26Re	AC, PS	5650	>2700	>3400	>3000
W-0.5Cb	AC, PS	6160	2[illegible]00-3000	3100-3400	2700-2900

(a) PS - pressed and sintered; AC - arc cast; EB - electron-beam melted, PAS - plasma-arc sprayed and sintered.

Tungsten and tungsten-alloy products historically have been made by powder-metallurgy techniques, and these are still the primary means for commercial production of bar and billet stock. The most reliable of these methods for producing forgeable tungsten alloys is isostatic pressing followed by sintering in vacuum or hydrogen. Pressed and sintered tungsten billets normally exhibit some porosity, particularly at the billet centers, and have a density less than theoretical. The amount of porosity increases with increasing section size, thus, the larger the billet, the lower the average density. Average densities exceeding 94 per cent ensure at least 90 per cent in all regions, but this is difficult to accomplish in billets larger than about 5-6 inches in diameter.

To overcome billet section-size limitations, larger parts have been produced by pressing and sintering a preform shape, and then hot forging to improve both the density and mechanical properties. Another method consists of plasma-arc spraying a preform shape onto a mandrel.[5] The mandrel is then removed and the blank sintered to a forgeable preform. Forging blanks prepared in this manner have been successfully forged into hollow conical shapes weighing up to 80 pounds.

Cast ingots of unalloyed tungsten and tungsten alloys can be produced by either consumable-electrode arc melting, electron-beam melting, or arc-melt centrifugal casting. The application of these techniques to the production of tungsten bar and billet stock is fairly recent in comparison with powder-metallurgy methods. Unalloyed tungsten ingots as large as 9 inches in diameter have been made by consumable-electrode arc melting, and ingots of binary W-Mo alloys such as W-15Mo have been melted in sizes up to 12 inches in diameter.[6] Electron-beam melting of unalloyed tungsten ingots up to about 4 inches in diameter has been accomplished, and vacuum-arc centrifugal casting of W-2Mo ingots up to about 6 inches in diameter has been reported.[4,6] By extrusion breakdown of the cast ingots, wrought forging billet stock ranging from about 1-1/2 to 5 inches in diameter can be produced.

FORGING BEHAVIOR

Like molybdenum, the forgeability of tungsten and tungsten alloys is directly related to billet density, grain size, and interstitial content and, therefore, to the method of billet production. The general influences of billet processing history on the forgeability of tungsten materials are summarized in Table 21-2.

The forgeability of pressed and sintered tungsten increases with increasing density. Production experience indicates that the minimum density for reasonable forgeability should be at least 90 per cent of theoretical. As mentioned above, this requires an average billet density of about 94 per cent. At this density level, pressed and sintered billets and preforms can be readily forged in all directions, i.e., upset, side forged, and roll forged.

Although there are indications that cast ingots of tungsten with a fine grain structure may be directly forged, the most reliable method of producing forgeable billet stock is by extrusion breakdown of the cast ingot. Based on work at Thompson-Ramo-Wooldridge, extruded billet stock is readily forgeable in either the stress-relieved or recrystallized condition provided the structure is fine grained and uniform.[4]

TABLE 21-2. INFLUENCE OF BILLET HISTORY ON FORGEABILITY OF TUNGSTEN AND TUNGSTEN ALLOYS

Billet Condition	Comments on Forgeability
Pressed and sintered	Billets up to about 6 inches in diameter can be forged in any direction. Since billet centers become weaker with increasing diameters, larger billets usually must be either extruded or side-forged before they can be upset by forging.
Slip cast and sintered	Not forgeable.
Arc cast	Usually not forgeable.
Arc cast and extruded	Forgeability increases with increasing extrusion ratio. A ratio of about 4:1 is about minimum for reasonable forgeability; can be both side and upset forged.
Electron-beam melted	Limited data indicate that as-cast tungsten can be directly forged if grain size is fine. Otherwise, ingots are usually not forgeable.

The levels at which interstitials begin to impair forgeability are not only a function of the total content, but also of the amount of grain surface available to accommodate them. This point is reflected in the superior forgeability of fine-grain, powder-metallurgy-tungsten compared with that of coarse-grain, arc-cast tungsten. Also, the critical transitions between ductile and brittle behavior of the latter are far more sensitive to small changes in interstitial composition. Consistently good forgeability is obtained in tungsten containing less than 75 ppm oxygen and less than 30 ppm nitrogen. A tentative Aerospace Materials Specification for pressed and sintered unalloyed tungsten forgings recommends the following maximum compositional limits for satisfactory forgeability and forging quality:(7)

Carbon	60 ppm
Oxygen	60 ppm
Nitrogen	20 ppm
Metallic elements, each (Al, Fe, Cu, Ni, Si)	20 ppm
Hydrogen	10 ppm
Others, each	10 ppm

Pilot-production forging of a thin-section tungsten shape in the 50-pound range indicates similar limits are applicable to arc-cast material.(4)

Tungsten and several W-Mo alloys have been successfully forged over a wide range of temperatures from as low as 1800 F to 3500 F. Temperatures in the vicinity of 3000 F are favored for initial breakdown of pressed and sintered billets, but finish forging is usually done at temperatures ranging from 2600 F downward to 2000 F. In

TABLE 21-3. TENSILE PROPERTIES OF SELECTED TUNGSTEN NOZZLE AND RING PARTS PRODUCED FROM PRESSED AND SINTERED BILLETS BY HOT-COLD WORKING TECHNIQUES[8]

Fabrication Method	Test Temperature, F	Axial Direction: Ultimate Strength, ksi	Axial Direction: Elongation, % in 1 inch	Axial Direction: Reduction of Area, %	Tangential Direction: Ultimate Strength, ksi	Tangential Direction: Elongation, % in 1 inch	Tangential Direction: Reduction of Area, %
Upset forged to 80% reduction, warm formed	RT	163.5	0.3	0.0	157.5	0.5	0.0
	200	162.8	6.0	2.0	148.5	3.5	0.0
	400	120.0	13.0	30.0	121.0	5.0	4.0
Ring rolled with 50% wall reduction (14-1/2" OD x 12" ID x 3-1/2")	300	100.5	0.0	0.0	126.0	2.0	0.0
	400	90.5	2.0	3.0	117.7	18.0	21.0
	500	82.0	14.0	18.0	111.8	32.5	55.5
Back extruded and ring rolled	400	113.9	10.6	40.3	111.5	27.5	56.7
Die forged to dogbone shape and ring rolled	400	110.9	32.0	56.1	118.1	18.5	45.2

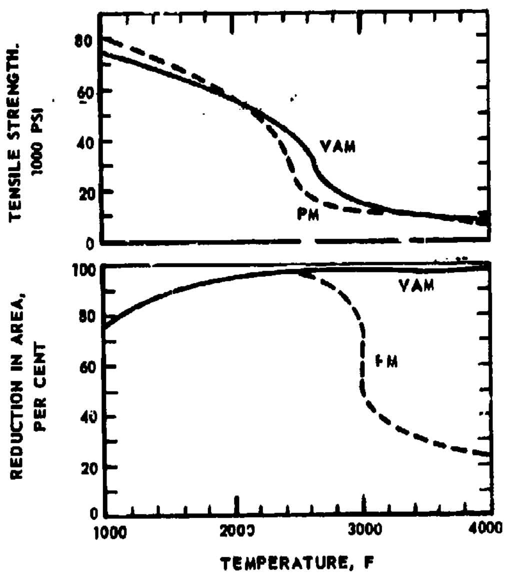


FIGURE 21-1. CHARACTERISTIC AS-WROUGHT TENSILE PROPERTIES OF VACUUM-ARC-MELTED AND POWDER-METALLURGY TUNGSTEN[9]

rocket-nozzle forging, a billet-height reduction of approximately 80 per cent at forging temperatures at all times below the critical recrystallization temperature is considered essential.[8] This imparts the fine-grained, fully wrought, fibrous structure that is characteristic of high strength and good ductility at low transition temperatures, as indicated by the data in Table 21-3.

Similar working temperature ranges appear to be applicable to arc-cast unalloyed tungsten, W-2Mo, and W-15Mo, except that breakdown operations are performed by extrusion rather than forging. Optimum forgeability of arc-cast and extruded billets of these three materials is obtained at temperatures ranging downward from about 2350 F to 2100 F, according to Lake.[4] This is also coincident with achieving the most desirable structure in the forging from the standpoint of properties.

Recrystallization temperatures for tungsten materials vary from about 2500 F to 3200 F, depending on the amount of prior deformation. When forged below its recrystallization temperature tungsten work hardens; thus, progressively higher forging pressures are required with increasing reduction and decreasing temperature. A measure of the relative forging-pressure requirements and the forgeability at various forging temperatures can be obtained, respectively, from elevated-temperature tensile strength and ductility data. As illustrated by the data for unalloyed tungsten in Figure 21-1[9], the strength increases quite drastically as the temperature is reduced from 3000 F to 2000 F. Similarly, the ductility of pressed and sintered tungsten improves as the temperature is reduced below 3000 F. The ductility of vacuum arc melted, on the other hand, is not very sensitive to temperature variations in this range. The tensile strength of tungsten is compared with that for various other types of alloys at their respective forging temperatures in Table 21-4. The higher strength of tungsten is indicative of higher forging pressure requirements.

TABLE 21-4. TENSILE STRENGTHS OF SEVERAL ALLOYS AT THEIR RESPECTIVE FORGING TEMPERATURES[2]

Metal or Alloy	Forging Temperature, F	Ultimate Strength, psi
AZ80 Magnesium	750	3,500 - 4,000
6061 Aluminum	800	4,000 - 5,000
304 Stainless Steel	2100	5,000 - 7,000
Ti-6Al-4V	1700	10,000 - 15,000
Waspaloy (Ni-base)	2000	10,000 - 15,000
Mo-0.5Ti-0.08Zr[a]	2500	20,000 - 30,000
Unalloyed W[a]	2500	25,000 - 40,000

(a) Data represents recrystallized condition.

Upset forging studies at Thompson-Ramo-Wooldridge[4] showed that initial deformation pressures for the W-15Mo alloy at temperatures between 2000 F and 3000 F vary between about 90,000 and 105,000 psi. Upon reaching 50 per cent upset reduction, the forging pressures were estimated to be in the vicinity of 150,000 psi. A significant point is that a 1000 degree F change in temperature had little influence on forging-pressure requirements. This corresponds well with the forging-pressure behavior of the molybdenum alloy Mo-0.5Ti-0.08Zr over this temperature range.[3] The initial upset forging pressure drops off appreciably on heating above 3000 F, as shown in Figure 21-2, but rupturing occurs in increasing degrees of severity as the temperature increases. In the range from 3000 F to 4000 F, the initial upsetting pressures for the W-15Mo alloy are twice those for the Mo-0.5Ti-0.08Zr alloy, and nearly five times those for unalloyed molybdenum.[3,10]

Similar data for W-5Mo and W-0.5Cb alloys are included in Figure 21-2. Both of these alloys require higher initial forging pressures than W-15Mo at temperatures below about 2500 F, but their pressures drop off more rapidly at higher temperatures. These data reflect the differences in recrystallization behavior of the alloys. The forging loads for 50 per cent upset reductions for the three alloys also reflect this behavior as a function of temperature, as shown in Figure 21-3.

COMMENTARY ON FORGING PRACTICES

Metallurgical principles in the forging of tungsten are much the same as those for molybdenum. Tungsten is generally forged in the hot-cold-work temperature range where hardness and strength increase with increasing reductions. Both systems exhibit increasing forgeability with decreasing grain size.

Tungsten requires considerably higher forging pressures, which tax the stress limits of most forging die materials. For this reason, it is often necessary to use in-process recrystallization annealing treatments to reduce load requirements for subsequent forging steps. The need for lateral support during forging is greater for tungsten than for molybdenum, and the design of preliminary forging tools is more critical. This is particularly true for pressed and sintered billet stock which has some porosity and less than theoretical density.

Control of Reduction. Control of the amount of reduction imparted to tungsten during forging is quite critical. Initial reductions are generally limited to 30-40 per cent. This practice avoids excess hardening that may lead to fracture. Forgings are sometimes recrystallized at this point to achieve grain refinement, to lower subsequent forging-load requirements and, in the case of pressed and sintered billets, to further aid densification. Once complete densification is achieved, intermediate stress-relief annealing between successive forging operations is often adequate for good forgeability.

Opposing conical dies are sometimes used to provide lateral restraint to the billets during initial upsetting. This helps to minimize the possibility of center cracking of billets with height/diameter ratios of 1:1 or higher. The control of reductions for upset forging billets with small height/diameter ratios (e.g., 1:2) is not critical because the stresses are essentially compressive at the outset of deformation. Billets 5-1/2 inches in diameter by 3 inches high of pressed and sintered tungsten have been upset forged successfully to reductions exceeding 80 per cent.

In hot-cold forging a rocket nozzle shape from arc-cast and extruded billets of W-2Mo alloy and unalloyed tungsten, Lake[4] reported total upset reductions of 70 per cent were readily attainable by multiple upsets with intermediate stress-relief anneals. At 2350 F with billets having a height/diameter ratio of 1.75:1, the maximum initial upset reduction without sidewall restraint was limited to about 40 per cent to avoid shallow intergranular cracking on the sidewall. The complete forging sequence for producing a thin-walled unalloyed tungsten nozzle shape with a major diameter of 8 inches from arc-cast and extruded billet stock is shown in Figure 21-4. This part was forged on an 8000-ton mechanical crank press at the following temperatures for each operation:

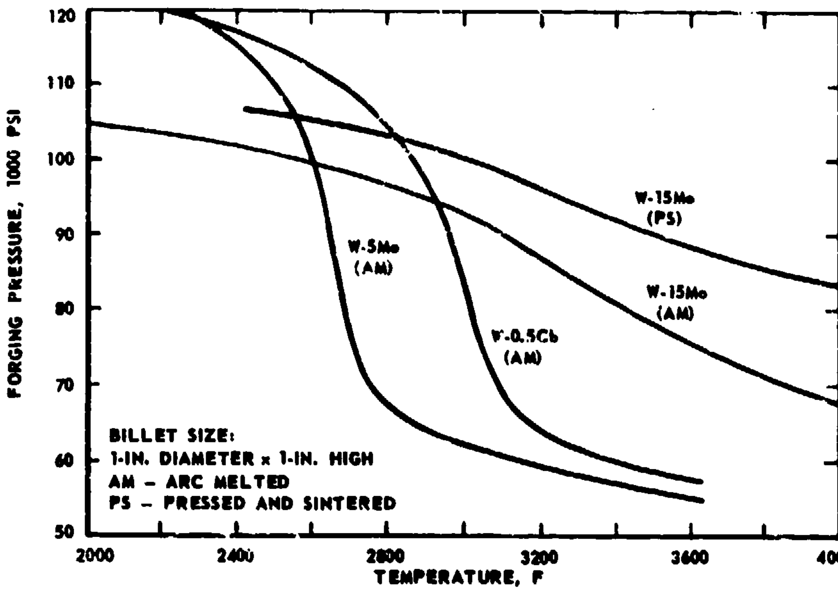


FIGURE 21-2. INITIAL UPSET PRESSURES FOR FORGING OF EXTRUDED BILLETS OF THREE TUNGSTEN ALLOYS AT VARIOUS TEMPERATURES After Lake(4)

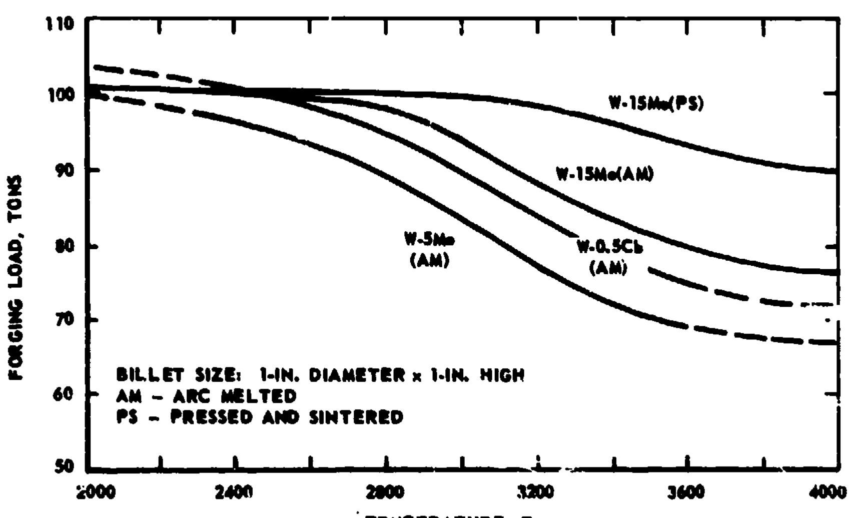


FIGURE 21-3. FORCE REQUIREMENTS FOR 50% UPSET FORGING OF EXTRUDED BILLETS OF THREE TUNGSTEN ALLOYS AT VARIOUS TEMPERATURES

After Lake(4)

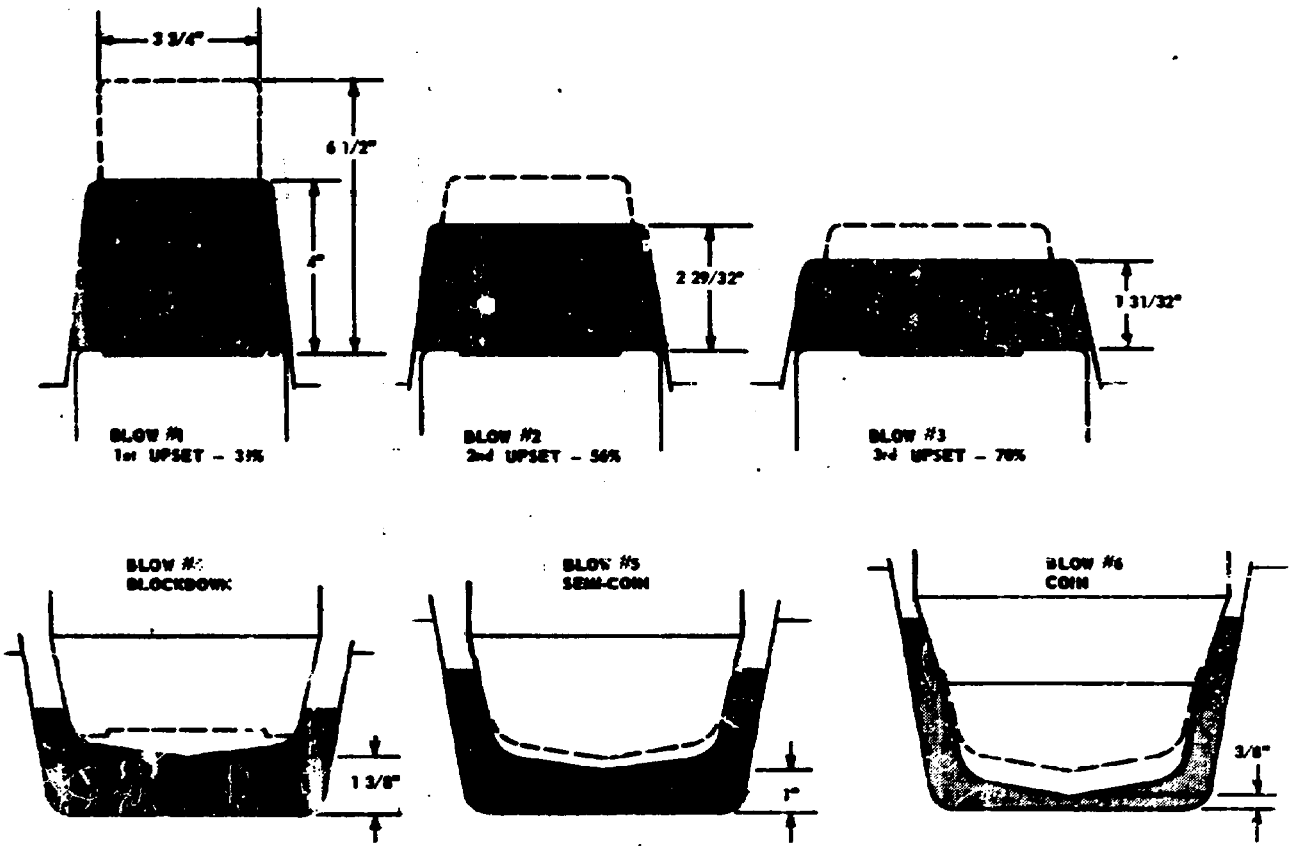


FIGURE 21-4. FORGING SEQUENCE FOR THIN-SECTION ROCKET-NOZZLE CONFIGURATION OF UNALLOYED TUNGSTEN[4]

Operation	Temperature, F
Upset No. 1	2350
Upset No. 2	2300
Upset No. 3	2300
Blockdown	2250
Semi-coin	2200
Coin	2200

Billet Heating. Like molybdenum, tungsten forms a volatile oxide at forging temperatures, and surface contamination is usually not a problem. Thus, except for the problem of creating undesirable vapors, conventional billet-heating furnaces are generally applicable for heating tungsten to temperatures up to about 2500 F. To reduce the amount of oxidation (and the metal loss), various types of glass coatings can be used on the billets. Also, in the case of electric resistance or induction furnaces, simply purging the furnace retort with inert gas is effective.

Die Materials. The selection of die materials for forging tungsten depends a great deal on the forging equipment being used. For hammer-forging dies, the common low-alloy Cr-Mo-V die steels are used with considerable success. The AISI H-11 and H-12 types of tool steel are often favored for press forging. Lake(4) reported excellent performance for dies of H-12 hardened to R_C 50-52 in crank press forging of tungsten and several tungsten alloys at 2100-2350 F; in forging a lot of 32 parts, die wear occurred only on the flash radii. It was estimated that at least 100 forgings could have been produced before polishing of the flash throat would have resulted in radii so large that trimming of flash from the forgings would become difficult.

Lubrication. Tungsten oxide, which becomes molten and volatilizes at forging temperatures, serves as an effective lubricant for forging "bare" tungsten. Mixtures of graphite and molybdenum disulfide sprayed on the dies aid in lubrication and act as effective parting agents to ease removal of the parts from the dies. Surface finishes of less than 100 microinches, rms, on small rib-and-web type parts are attainable by this practice.(4)

A variety of glasses that can be used as coatings to reduce oxidation are also effective as lubricants. Glasses have the added advantage of providing insulation between the dies and the much hotter workpiece, but rougher surfaces are produced. Also, there is a problem of glass collecting in the crevices of impression dies, which can interfere with achieving complete die fill. Thus, glasses are most applicable for simpler forging operations such as upsetting or back extrusion.

Cooling From Forging. Having a comparatively high ductile-brittle transition temperature, tungsten is subject to thermal cracking if cooled too rapidly below about

1000 F. For this reason, forgings are generally cooled slowly either in furnaces or by burying them in insulating materials after forging.

Since finish forging is almost always done in the hot-cold working range, a desirable practice is to stress relieve the parts directly after forging prior to slow cooling to room temperature.

COMMENTARY ON FORGING DESIGN

The greatest use for tungsten forgings has been in rocket-nozzle applications. Production forging of tungsten nozzle inserts has been largely limited to billets prepared by powder-metallurgy techniques because of their earlier availability than cast tungsten.

Until recently, the closed-die forging of tungsten was confined to generously contoured, solid shapes. Even slightly hollowed cones required back extrusion that often led to cracking. Thus, size capability was restricted to short, conical shapes in the vicinity of 10 inches in diameter and ring shapes up to about 15 inches in diameter. This problem has been largely overcome through improved billet uniformity and the development of improved forging practices. By careful die staging and by combined methods of fabrication such as upsetting, back extrusion, and ring rolling, size capability has been increased and thinner sections with improved tolerances and surface finish are possible. For example, Evans[8] reported the successful fabrication of a 33-inch-OD x 31-1/2-inch-ID x 4-inch-high r[illegible], weighing approximately 220 pounds from a pressed and sintered tungsten preform.

The production of conical tungsten shapes of the type required for rocket-nozzle applications, can be conveniently categorized into the three forging designs illustrated in Figure 21-5. Assuming the starting material is pressed and sintered billet stock with a nominal diameter of 5-1/2 inches, the general features of these forging designs would be roughly as follows.

<u>Type A. Solid Design</u>. This design would require a 5-1/2-inch-diameter x 15-inch long billet weighing about 250 pounds. The billet would be upset forged to about a 7-1/2-inch-thick biscuit between flat dies (about 50 per cent reduction). Because the length-to-diameter ratio of the billet would be nearly 3:1, initial upsetting would be done in small increments to avoid buckling. Upsetting would probably require at least one, and possibly two reheating steps. The upset biscuit would then be forged in the finishing die. A major disadvantage of this forging design is that the deformation imparted would be comparatively small and nonuniform. The metal nearest the top die would receive the greatest reduction while the metal near the bottom of the lower die would be essentially unworked.

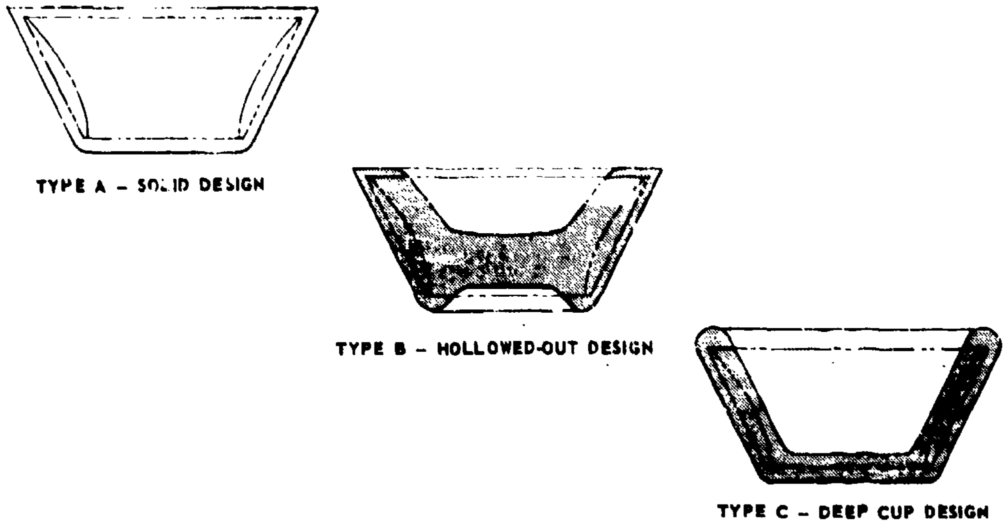
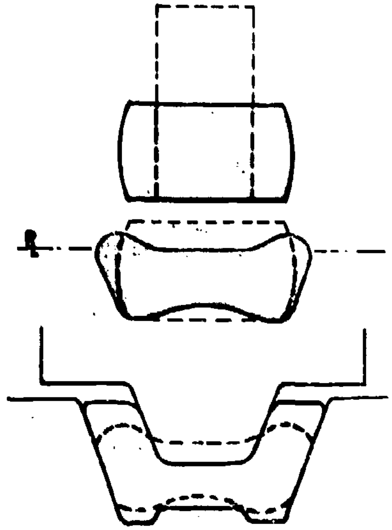


FIGURE 21-5. THREE TYPES OF FORGING DESIGNS FOR PRODUCING TUNGSTEN ROCKET NOZZLE SHAPES

Typical configuration of nozzle-throat insert is shown by dashed outline

Type B. Hollowed-Out Design. The typical forging sequences for this design are illustrated in Figure 21-6. The first step would be to upset a 5-1/2-inch-diameter x 11-3/4-inch-long billet weighing about 180 pounds to a biscuit about 6 inches thick, probably with one reheating. The second step would consist of forging the biscuit in a closed die to a "dog-bone" shape something like that illustrated. This die would be designed so that the side wall of this shape would be supported by the finishing die. Thus, the finishing die would provide lateral support to the part at the outset of deformation. Depending on the levels of deformation imparted, it might be desirable to recrystallize the dog-bone preform before finishing to reduce forging-pressure requirements. This design would require less material and the forging would receive greater and more uniform deformation than the Type A design. This type of forging design is quite versatile, and would be applicable for shapes other than simple cones.

Type C. Deep-Cup Design. This type of design is particularly suited to short conical shapes. The typical forging sequences for this design are illustrated in Figure 21-7. A billet about 5-1/2 inches in diameter and 8 inches high weighing about 140 pounds would be required. The first step would consist of upset forging to a biscuit about 1 inch thick (88 per cent reduction). It is likely that at least two, and possibly three, reheating steps would be required to accomplish the upset. The



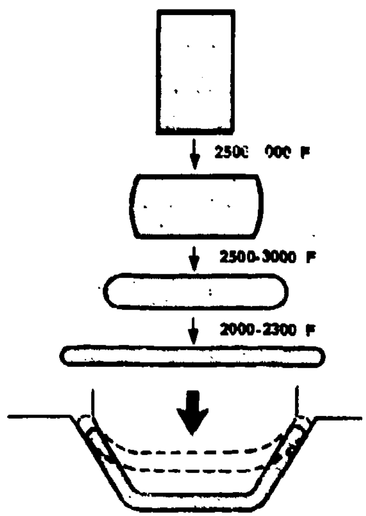
(1) UPSET BILLET TO FLUSH BISCUIT AT 2500-3000 F

(2) DIE FORGE TO "DOG-BONE" SHAPE SIZED TO FINISH DIES AT 2000-2300 F

(3) FORGE TO FINISH SHAPE AT 2000-2300 F

FIGURE 21-6. PROBABLE DIE STAGES IN FORGING PRESSED AND SINTERED TUNGSTEN FOR TYPE B FORGING DESIGN

Blanks would be reheated in each stage as needed.



(1) UPSET BILLET TO ABOUT 1-INCH PANCAKE IN MULTIPLE REDUCTIONS OF ABOUT 50%. RECRYSTALLIZE PRIOR TO FINAL UPSET

(2) DRAW TO FINISH SHAPE IN PROGRESSIVE STAGES AT ABOUT 2000-2300 F

FIGURE 21-7. PROBABLE DIE STAGES IN FORGING PRESSED AND SINTERED TUNGSTEN FOR TYPE C FORGING DESIGN

1-inch-thick plate would then be cupped by a hot-drawing operation. To lower forging pressure requirements, the biscuit might best be recrystallized before the final upsetting stage, which would restore the desired wrought structure for cupping. Quite likely the drawing operation illustrated would be too severe to avoid rupturing of the blank, and a multistage drawing operation would be required. A single draw operation probably would be limited to shallower cones less than 4 inches high, having diameter-to height ratios exceeding 2 to 1. An important feature of this design approach is that the part receives more uniform deformation than do those with Design Types A or B.

Another method for forging tungsten into slightly tapered conical shapes of Type C design is by back extrusion. A typical sequence for such parts was illustrated in Figure 21-4. The procedure consists of upsetting to a biscuit that fits the container and then back extruding around a punch. Following this technique with arc-cast and extruded billets of unalloyed tungsten and W-2Mo alloy, Thompson-Ramo-Wooldridge[4] produced pilot lots of experimental thin-wall forgings with very close tolerances, as shown in Figure 21-8. The properties obtained on the forged parts are given in Table 21-5.

TABLE 21-5. TENSILE PROPERTIES OF THIN-WALL TYPE C SHAPES FORGED FROM ARC-CAST TUNGSTEN AND W-2Mo ALLOY

Test Temperature, F	Ultimate Strength, ksi	0.2% Yield Strength, ksi	Reduction of Area, %	Elongation, % in 1 inch
	W-2Mo Alloy - As-Extruded Billet Stock			
310	161.2	160.5	1	4.8
400	144.3	143.3	12	15.1
540	121.8	121.3	27	13.8
655	117.4	116.9	30	10.6
3000	52.4	47.1	79	22.1
3900	8.3	6.2	99	91.0
	Unalloyed W - Recrystallized Billet Stock			
200	166.4	163.8	2	1.4
300	144.4	141.0	5	10.5
350	132.9	130.0	37	19.4
450	114.6	110.2	51	[illegible]
3000	16.2	7.3	85	66.0
3500	11.5	6.0	96	77.8
	Unalloyed W - As-Extruded Billet Stock			
200	167.7	167.5	3	3.1
300	147.3	146.3	16	[illegible]
350	137.5	136.7	13	[illegible]
400	119.3	119.3	30	17.[illegible]
3000	15.8	8.4	98	69.0
3500	10.4	5.6	98	64.[illegible]

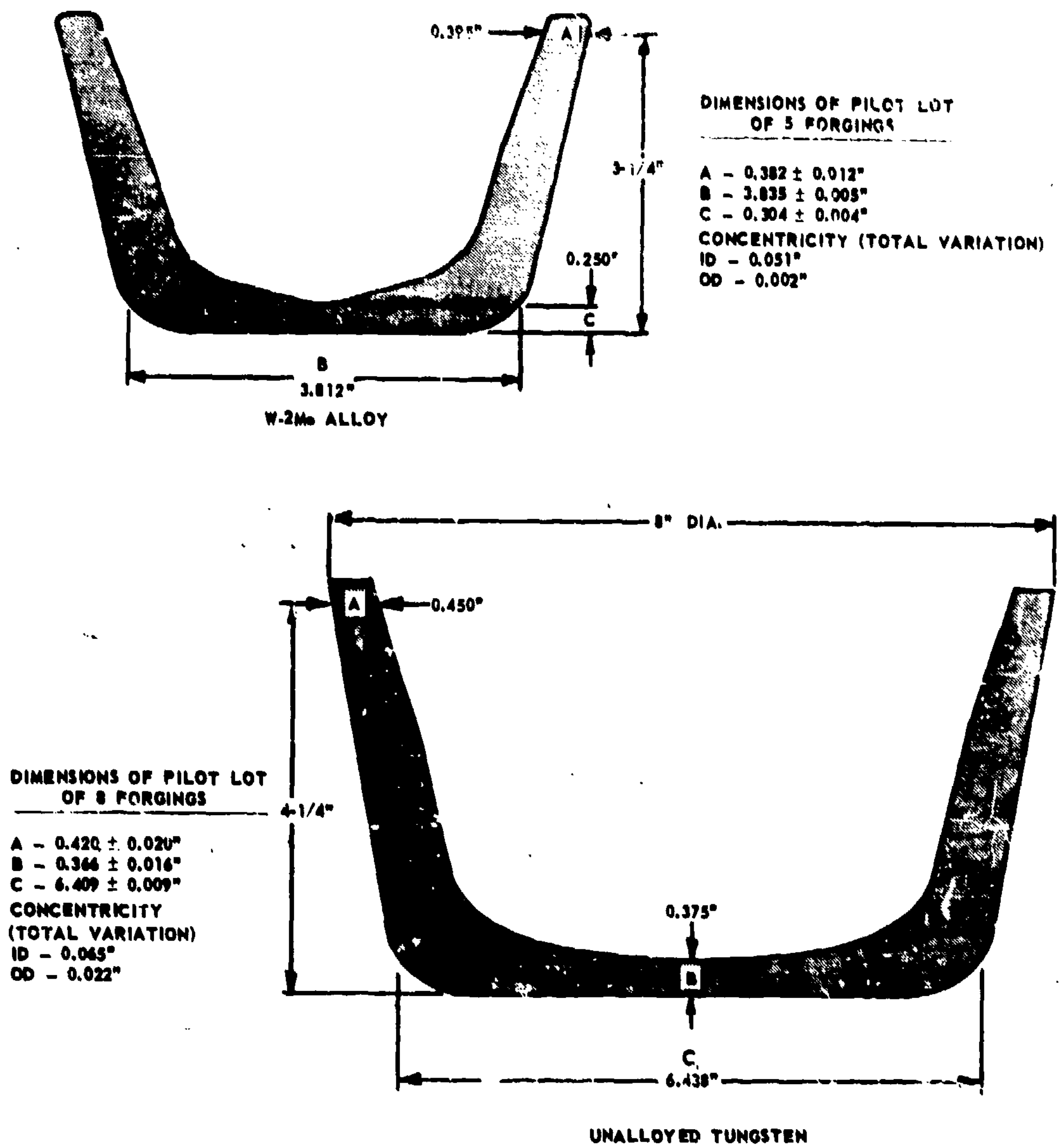
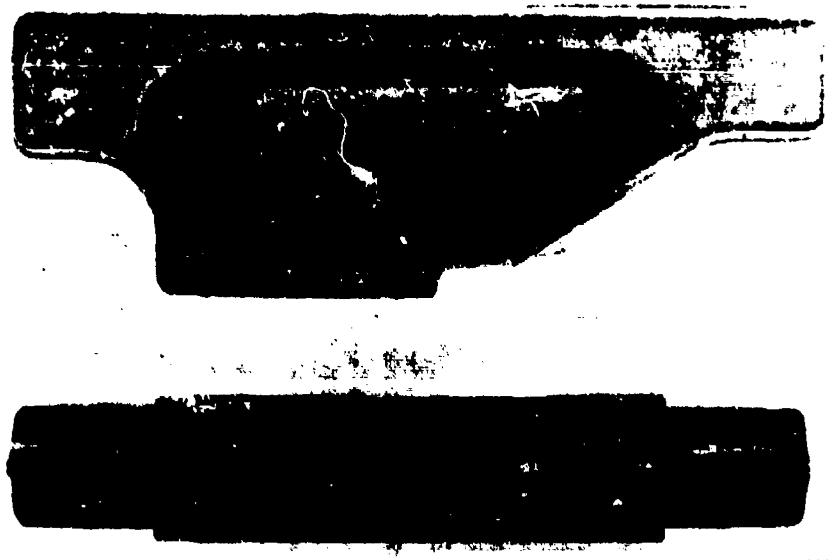


FIGURE 21.8. NOMINAL DESIGN AND ACTUAL DIMENSIONS OF TWO SIZES OF TYPE SHAPES OF TUNGSTEN FORGED FROM ARC-CAST AND EXTRUDED BILLETS(4)



N90332

FIGURE 21-9. STRUCTURAL PART OF W-15Mo ALLOY FORGED TO PRECISION TOLERANCES

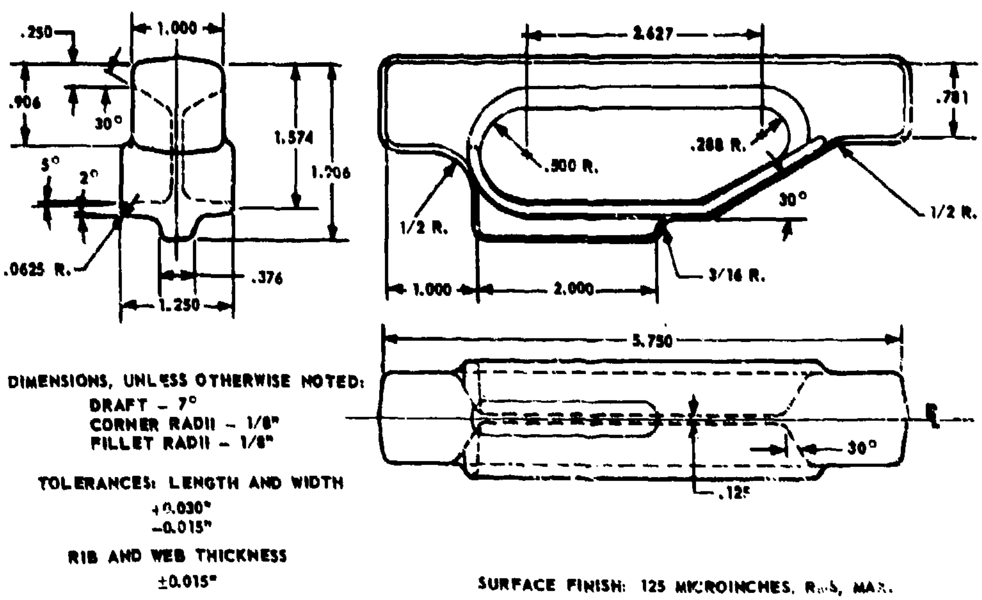


FIGURE 21-10. DETAIL DIMENSIONS AND TOLERANCES FOR EXPERIMENTAL PRECISION FORGING OF W-15Mo ALLOY

An experimental structural part of W-15Mo alloy forged to precision design tolerances is shown in Figure 21-9.[4] This part was forged on a 1500-ton mechanical crank press in three operations, starting with a 1-1/8-inch-diameter x 5-1/4-inch-long billet:

(1) Preblock to a dog-bone shape at 2100 F

(2) Block down to generously contoured part outline at 2150 F

(3) Coin to finish size at 2150 F.

The parts were sand blasted, vacuum stress relieved at 2400 F, and trimmed of flash after each operation. Details of the part design are shown in Figure 21-10. In a pilot-production run of 32 billets, 31 sound forgings were produced with total dimensional variance less than that allowed by the precision-design tolerances. This experimental program demonstrated the feasibility of applying conventional forging practices to the production of tungsten shapes with close dimensional control and good surface finishes from arc-cast and extruded billet stock.

REFERENCES

(1) Schmidt, F. F., and Ogden, H. R., "The Engineering Properties of Tungsten and Tungsten Alloys", DMIC Report 191, Defense Metals Information Center, Battelle Memorial Institute (September 1963).

(2) Henning, H. J., and Boulger, F. W., "Notes on the Forging of Refractory Metals", DMIC Memorandum 143, Defense Metals Information Center, Battelle Memorial Institute (December 21, 1961).

(3) Henning, H. J., Sabroff, A. M., and Boulger, F. W., "A Study of Forging Variables", Technical Documentary Report ML-TDR-64-95, Contract AF 33(600)-42963, Battelle Memorial Institute (April 1964).

(4) Lake, F. N., "Tungsten Forging Development Program", Technical Documentary Report ML-TDR-64-128, Contract AF 33(600)-41629, Thompson Ramo Wooldridge, Inc. (April 1964).

(5) Singleton, R. H., Bolin, E. L., and Carl, F. W., "The Fabrication of Complicated Refractory Metal Shapes by Arc-Spray Techniques", High Temperature Materials, pp 641-654, Interscience, Cleveland, Ohio (1961).

(6) Riesen, A. E., Wang, C. T., and Worcester, S. A., "Tungsten Extrusion Development Program", Technical Documentary Report ML-TDR-64-2[illegible], Contract No. AF 33(600)-42395, Wah Chang Corporation (May 1964).

(7) "Tungsten Forgings", Aerospace Material Document No. AMD 78BU Society of Automotive Engineers, New York City (January 15, 1963).

(8) Evans, F. D., "Forging of Tungsten Components", paper presented at the Technical Conference on Applied Aspects of Refractory Metals, sponsored by the AIME, Los Angeles, California (December 9-10, 1963).

(9) Ratliff, T. L., and Ogden, H. R., "A Compilation of the Tensile Properties of Tungsten", DMIC Memorandum 157, Defense Metals Information Center, Battelle Memorial Institute (September 11, 1962).

(10) Tombaugh, R. W., et al, "Development of Optimum Methods for the Primary Working of Refractory Metals", WADD Technical Report 60-418, Part II, Contract No. AF 33(616)-6377, Harvey Aluminum, Inc. (August, 1961).

CHAPTER 22

BERYLLIUM

Crystalline Structure:	Close-packed hexagonal
Crystalline Changes:	CPH to BCC on heating above 2300 F
Number of Phases:	One at forging temperature
Liquidus Temperature:	2345 F
Solidus Temperature:	Over 2300 F
Reaction When Heated in Air:	Oxidation above about 1400 F

GRADES AND FORMS AVAILABLE

Several common grades of beryllium, their basic compositions, and the forms in which they are available for forging are listed in Table 22-1. These beryllium grades which contain BeO in nominal amounts ranging from about 0.8 to about 4.3 weight per cent may be considered as alloys of beryllium, in that they have been shown to age harden and possess other characteristics of alloys.(1) Otherwise, the nearest approach to a commercial beryllium-base alloy system would be the more recent beryllium-aluminum alloys containing from about 24 to 43 per cent aluminum.(2)

TABLE 22-1. COMPOSITIONS OF SEVERAL GRADES OF BERYLLIUM AVAILABLE FOR FORGING

	Nominal Composition, wt %								
Grade	Be min	BeO max	Al max	C max	Fe max	Si max	Mg max	Others max	Forms Available(a)
Nuclear	99.0	0.9	0.075	0.10	0.10	0.06	0.08	Trace	P, B
Structural	99.0	1.0	0.10	0.12	0.15	0.08	0.08	0.04	P, B
	98.5	1.2	0.14	0.15	0.16	0.08	0.08	0.04	P, B
	98.0	2.0	0.16	0.15	0.18	0.08	0.08	0.04	P, B
	97.4	3.0	0.18	0.20	0.20	0.12	0.08	0.06	P, B
Instrument	92.0	4.25 (min)	0.20	0.50	0.50	0.15	0.10	0.10	B

(a) P = powder, B = hot-pressed block.

Beryllium exhibits many desirable engineering characteristics: high strength, high modulus of elasticity, low density, and oxidation resistance up to about 1400 F. However, the metal has seen only limited applications because it is expensive, brittle, and toxic when inhaled in the metal powder form.

Beryllium billet stock can be produced either by consumable-electrode vacuum arc melting or by vacuum hot pressing of powder. Because hot-pressed block is more readily workable than cast ingot, it has been the principal form of billet stock used for forging. Beryllium block in sizes up to 6 feet in diameter and weighing up to about 5 tons has been produced. Billet manufacture can be bypassed by canning loose beryllium powder in evacuated steel containers and hot forging to consolidate the powders during deformation. This technique, described in detail later, has been used to produce a variety of forged parts weighing from about a pound to over a ton.(3)

FORGING BEHAVIOR

Beryllium, in many respects, behaves like magnesium during plastic deformation. Both metals are characterized by a CPH crystalline structure, which imparts low ductility at room temperature. The plasticity of beryllium is dependent on grain size (brittle when coarse, more ductile when fine). Because it is strain-rate sensitive, it exhibits better forgeability in presses than in hammers. Also like magnesium, beryllium exhibits strongly anisotropic flow characteristics. Essentially, this means that once flow has begun in one direction, extensive plastic flow in that direction can occur. This increasing plasticity in one direction is usually accompanied by reduced plasticity in other directions. (This effect might be termed "oriented plasticity".)

Since beryllium recrystallizes in the vicinity of 1400 F, it exhibits hot-working behavior above and cold-working behavior below this temperature. The similarities between magnesium and beryllium end when their ductilities in the hot-working range (above their respective recrystallization temperatures) are compared. Beryllium does not exhibit the sharp rise in ductility that magnesium does.

Beryllium has a low tolerance for cold work. Its ductility increases to some extent on heating, reaching the highest values at about 800 F and 1400 F.(4-7) Even at these temperatures, however, beryllium is generally less ductile than most other materials. For comparison, elongations reported for heated beryllium specimens are listed below for various materials tested in the vicinity of 800 F:

Material	Elongation, per cent
Hot-pressed block	30-40
Hot-pressed and extruded bar	15-20 (longitudinal) <5 (transverse)
Arc-cast and extruded bar	25 (longitudinal) <5 (transverse)

The low transverse ductility of the wrought material is characteristic of beryllium. One would expect poor upset ductility from such material, and this is the case with extruded bar.(8)

All forms of beryllium exhibit unstable structures when deformed above 1400 F. As little as 2 per cent deformation can promote recrystallization and subsequent grain growth, leading to wide variations in ductility.

Above 1400 F, it is extremely important to protect beryllium from oxidation and/or to incorporate suitable safety measures against possible beryllium oxide poisoning.* Because beryllium oxidizes at a slow rate up to 1400 F, it is safe to forge the material below 1400 F without resorting to elaborate safety measures against toxicity problems. Thus, much forging development work has centered around the use of temperatures between 1300 F and 1400 F for forging. In this range, grain growth is slowest and the metal exhibits mostly hot-working behavior.

In a program at the Ladish Company(8), upset forging trials were conducted on small billets produced by four techniques:

(1) Consumable-electrode vacuum arc melted

(2) Vacuum arc melted and hot extruded

(3) Vacuum hot pressed

(4) Vacuum hot pressed and hot extruded.

Forging trials were conducted over a range of temperatures, reductions, and speeds. All but a few samples were encased in 1/2-inch-thick mild steel jackets. Attempts to gain adequate support from thinner jackets (0.01 inch electroplated nickel + 1/16 inch mild steel) were unsuccessful. Important conclusions based on these upset forging trials were:

(1) Beryllium exhibits better forgeability when upset slowly in a hydraulic press than when upset rapidly in a drop hammer.

(2) Lateral support is required for upset reductions beyond about 25 per cent.

(3) Billets produced by vacuum hot pressing showed better forgeability in the as-pressed condition than when extruded. Samples were successfully forged with reductions exceeding 85 per cent. (The evaluation of arc-cast and arc-cast-and-extruded materials was hampered because of difficulties in obtaining sound, fine-grained billets.)

(4) Best forgeability was obtained by upset forging at pressing speeds, at temperatures between 1300 F and 1400 F, and at reductions less than 40 per cent.

(5) Forging pressures on the order of 10,000-12,000 psi were required for upsetting nickel-plated samples at temperatures from 1650 F to 2050 F. This indicates that the pressures required for forging beryllium are similar to those required for some of the magnesium alloys (e.g., AZ80 requires 11,000 psi at 600 F).

*References 16-19 contain information regarding the handling of beryllium oxide and many of the precautions necessary for various shop handling procedures.

(6) Highest strengths at room temperature and at 800 F were obtained by forging at temperatures between 1300 and 1450 F (tensile data are presented in Table 22-2).

In work at Wyman-Gordon reported by Cieslicki[3], studies were made to determine the influence of forging temperature on the mechanical properties of various shapes produced by the canned-powder technique. Comparisons were made with parts forged at 1600 F and 1900 F. As shown in Table 22-3, the 1600 F forging temperature imparts higher strengths and slightly higher average elongation values.

One of the features of beryllium shapes produced by the canned-powder technique is that the mechanical properties of the parts are less direction sensitive. This comes from the fact that there is comparatively little metal flow occurring in forging of beryllium powder. This, in turn, reduces the crystalline anisotropy ordinarily found in severely deformed wrought beryllium.

The most widely used production specification for commercial beryllium components forged from canned powders calls for the following mechanical properties:

Ultimate tensile strength	40,000 psi
0.2 per cent yield strength	30,000 psi
Elongation	1% in 1 inch

This specification can be readily met in production forging operations, as evidenced by the data in Table 22-4 for a production run of over 400 parts.[3] It will be noted, however, that both the strength and ductility of parts forged from hot-pressed block (standard structural grade containing about 2 per cent BeO) are substantially higher than parts forged by the canned-powder technique from an equivalent grade. One commercial specification[9] for forging from hot-pressed block of the 2 per cent BeO grade cites the following typical minimum properties:

Ultimate tensile strength	60,000 psi
0.2 per cent yield strength	35,000 psi
Elongation	3% in 1 inch

Compared with available data on forgings, these values are quite conservative.

According to Hornak and O'Rourke[10], significantly higher strengths at both room and elevated temperatures can be attained with instrument grade of beryllium containing over 4 per cent BeO. Comparative tensile properties on a simple wheel forging produced from standard commercial-grade (2 per cent BeO) and instrument-grade (<4 per cent BeO) hot-pressed block are presented in Table 22-5. It was observed in this work by The Brush Beryllium Company and Ladish Company, however, that this improvement in properties is gained at the expense of reduced forgeability.

TABLE 22-2. EFFECT OF FORGING TEMPERATURE ON TENSILE PROPERTIES OF HOT-PRESSED AND UPSET-FORGED BERYLLIUM[8]

Samples upset forged with reductions between 83 and 86 per cent.

Nominal Forging Temperature, F	Test Temperature, F	Average Ultimate Strength, ksi	Average 0.2% Yield Strength, ksi	Average Elongation, % in 1 inch	Average Reduction of Area, %
1300	80	78.6	[illegible]	13.5	14.2
1375	80	81.7	50.5	9.7	13.7
1450	80	77.0	42.5	16.0	14.0
1550	80	75.4	38.4	12.8	12.7
1650	80	69.5	32.8	9.3	6.7
1750	80	67.6	33.6	9.3	10.2
1900	80	63.8	33.3	7.5	8.1
2050	80	59.3	28.2	6.1	6.1
1300	800	32.3	30.8	21.6	48.4
1375	800	31.4	31.4	13.7	56.1
1450	800	31.2	27.7	30.8	63.1
1550	800	30.4	24.6	29.3	59.0

TABLE 22-3. ROOM-TEMPERATURE MECHANICAL PROPERTIES OF THREE PARTS FORGED AT 1600 F AND 1900 F BY THE CANNED-POWDER TECHNIQUE[3]

Forging Temperature, F	Part[(a)]	Average Ultimate Strength, ksi	Average 0.2% Yield Strength, ksi	Average Elongation, per cent
1600	A	67.0	55.0	2.00
	B	58.0	36.0	1.93
	C	60.9	41.1	2.90
1900	A	58.4	36.1	[illegible]
	B	43.9	32.8	1.86
	C	52.2	38.6	1.20

(a) Part descriptions:

Part A - 14-inch-long angle shape with 3-1/2-inch legs, 5/8 inch thick

Part B - 12-inch-long shaft having 3-inch-diameter x 3-inch-long hollow en... and 4 x 4-inch-square section 6 inches long at center. Ends flanged to 5 inch diameter

Part C - 8-inch-ID x 5-inch-ID x 4-inch-thick ring with three 3-inch-long run... extending from the OD.

TABLE 22-4. RANGE AND DISTRIBUTION OF ROOM TEMPERATURE MECHANICAL PROPERTIES FOR BERYLLIUM PARTS FORGED BY THE CANNED-POWDER TECHNIQUE[3]

Test bars removed from 441 parts having hollow bowl configuration.

Ultimate Strength, ksi	Number of Parts	0.2% Yield Strength, ksi	Number of Parts	Elongation, per cent	Number of Parts
<45	5	<30	1	<0.5	1
45-50	121	30-35	84	0.5-1.0	3
50-55	195	35-40*	189	1.0-2.0	111
55-60	45	40-45	131	2.0-3.0	100
60-65	36	45-50	27	3.0-4.0	69
65-70	8	50-55	5	4.0-5.0	34
>70	4	55-60	1	5.0-6.0	5
		>60	2	6.0-7.0	3
				>7.0	1

TABLE 22-5. COMPARATIVE PROPERTIES OF A WHEEL PART FORGED FROM STRUCTURAL (2% BeO) AND INSTRUMENT (>4% BeO) GRADES OF BERYLLIUM HOT-PRESSED BLOCK[10]

		Tensile Properties[a]		
Grade	Test Temperature, F	Ultimate Strength, ksi	0.2% Yield Strength, ksi	Elongation, % in 1 inch
Instrument	RT	110.0	75.0	10.0
Structural	RT	77.6	47.0	10.7
Instrument	500	81.1	67.2	13.7
Structural	500	54.6	43.1	52.0
Instrument	900	56.6	54.8	10.8
Structural	900	34.3	32.2	43.6
Instrument	1100	45.4	44.8	12.0
Structural	1100	28.7	28.6	28.5

(a) Both materials were forged in a similar manner with the same amount of reduction. Test specimens were taken at similar radial locations near the rim of the wheel.

While beryllium exhibits increasing strength with increasing reductions, it also exhibits increasingly pronounced anisotropy. Table 22-6 compares room-temperature tensile properties and the transverse/longitudinal strength ratios for circular forgings receiving differing amounts of reduction in differing directions of metal flow. A general aim in forging beryllium is to provide a minimum amount of metal flow consistent with the development of uniform mechanical properties part-to-part. This way, the ratio of transverse strength to longitudinal strength can be maintained closer to the desired value of 1.0.

TABLE 22-6. MECHANICAL PROPERTIES OF DIE FORGINGS SHOWING RELATIONSHIP BETWEEN AMOUNT AND DIRECTION OF METAL FLOW AND THE DEGREE OF ANISTROPY

Relative Amount of Reduction	Direction of Metal Flow (a)	Direction of Test (b)	Tensile Properties: 0.2% Yield Strength, ksi	Ultimate Strength, ksi	Elongation, % in 4D	Reduction of Area, %	Strength Ratio (c), U_T/U_L
1. Medium	R. A.	T	33-37	56-63	2.0-2.8	2.0-3.2	1.2
		L	34	46-54	1.1-2.3	1.6-2.4	
2. Medium	A	T	40-46	60-70	3.3-8.6	3.2-8.2	1.02
		L	34-38	62-64	3.7-7.2	4.7-7.1	
3. Medium	R	T	41	58-64	2.6-3.5	3.2-4.0	0.9
		L	41-44	67-68	4.6-6.9	4.7-7.1	
4. Large	A	T	54-55	54-58	0.4-1.8	0.6-1.7	0.8
		L	55-56	77-86	2.9-6.5	5.0-7.7	

(a) R - flow in radial direction; A - flow in axial direction; R. A. - flow in both radial and axial directions.
(b) T - transverse to direction of predominant flow; L - parallel to direction of predominant flow.
(c) Ratio of average transverse ultimate strength to average longitudinal ultimate strength.

The amount of deformation imparted during forging also influences the degree of preferred orientation. Samples upset forged from hot-pressed billet with about 85 per cent reduction show preferred orientation [determined by X-ray of the weakest basal (0001) plane], as high as 8 times random orientation.(8) Two die forgings, representing lower levels of reduction, exhibited preferred orientations up to 3.5 times random. These preferred orientation values are low compared with extrusions (18:1 extrusion ratio) having values of 14R and cross-rolled sheet having values as high as 28R.(11)

Thus, it is important to limit the amount of deformation during forging to avoid excessive preferred orientation. The actual limits for usable properties have not yet been established; however, it seems that reductions could be limited to levels of the order of about 60 per cent reduction.

COMMENTARY ON FORGING PRACTICES

Forging of beryllium, according to Hornak and O'Rourke[12], can be broken down into at least five significantly different forging techniques, as shown by the comparison in Table 22-7. The economic advantage of any one technique over another method would depend upon the defined characteristics of the desired end product. Thus, the choice of the forging technique would be governed by such factors as part size, tensile-property requirements, isotropic-property requirements, machining costs, etc.

TABLE 22-7. SUGGESTED CLASSIFICATION OF BERYLLIUM FORGING TECHNIQUES[12]

Forging Technique	Typical Forging Temperatures, F	Forging Pressure	Duration of Loading	Atmosphere	
				Preheat	Forging
Hot pressing[a]	1950	500 psi	hr-days	Vacuum	Vacuum
Forging canned powder	1600-1950	10-20 tsi	sec-min	Inert	Air
Forging steel-clad hot-pressed block	1400-1950	20-40 tsi	sec	Inert or air	Air
Forging bare hot-pressed block using support rings	1200-1500	30-60 tsi	sec	Air	Air
Forging bare hot-pressed block	1200-1500	30-60 tsi	sec	Air	Air

(a) Hot pressing is included for comparison and can be considered as a type of "creep" forging operation.

Based on the starting material rather than the forging procedure, a more general breakdown of beryllium forging would comprise only two basic categories:

(1) Canned-powder forging

(2) Billet forging.

By this classification, the processes are more distinct with respect to the forging practices and the production capabilities in terms of part geometry and properties.

Canned-Powder Forging

The canned-powder forging process is illustrated in its simplest form in Figure 22-1 for the case of an upset disk. Since the greatest density of powders is only about 50 per cent of theoretical, the container is designed to hold a little more than twice the volume of the fully dense forging. The container shape is ordinarily geometrically similar to the final part configuration, with allowances to provide for enough flow to fill the die. An example of the canned-powder billet shape and its relative size for producing an actual part (a flat-bottom cup) is shown in Figure 22-2 [12]

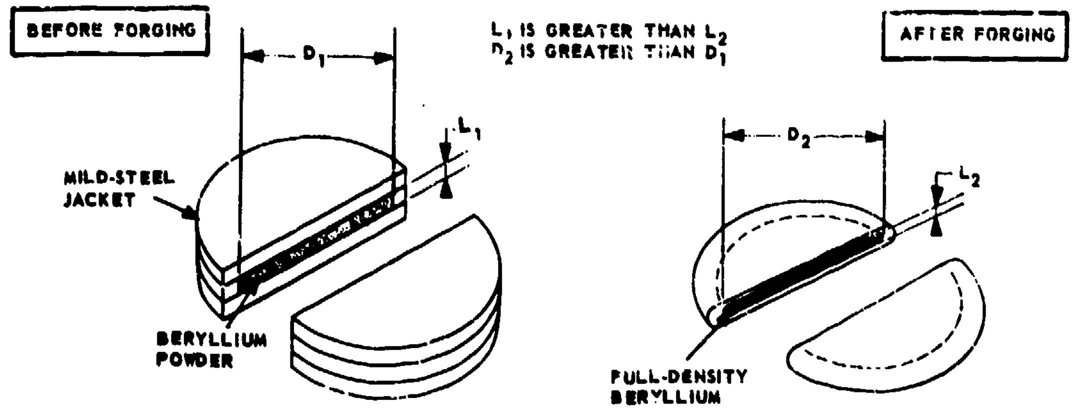


FIGURE 22-1. SCHEMATIC DIAGRAM OF UPSET FORGING BY CANNED-POWDER METHOD

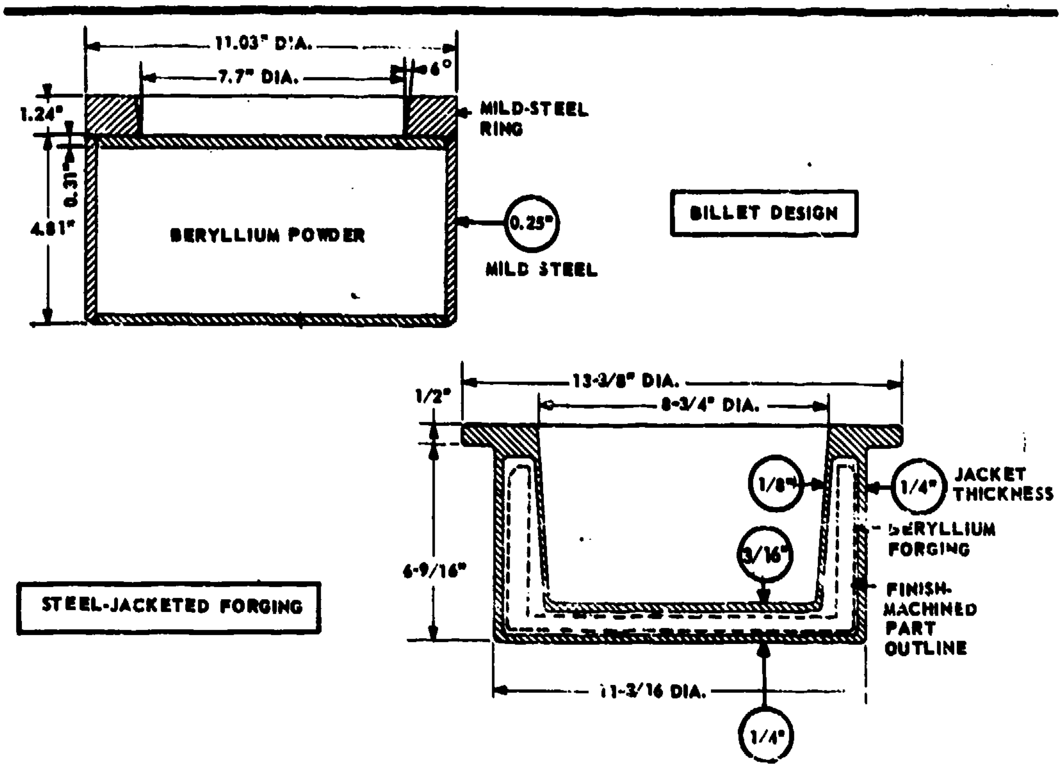


FIGURE 22-2. DETAILS OF CANNED-POWDER BILLET DESIGN FOR FORGING A BERYLLIUM CUP SHAPE(12)

When adequately supported by the dies, the container material may be thin, mild-steel sheet. When die support is not possible, the container walls are generally made from heavier plate to provide sufficient pressure on the powder during forging. Thick-walled plates of stainless steel (e.g., Type 304) are sometimes used to provide even greater constraint.

Powders are loaded through ports provided in the enclosed containers. Vibration is used to obtain the highest packing density of the powder. The containers are sometimes purged with argon before loading to reduce oxidation during subsequent heatup. The ports of the packed containers are then sealed off with threaded plugs. Unusually large containers are sometimes evacuated to reduce the volume of entrapped gases.

The sealed containers are heated in the temperature range of 1600 F to 1900 F for forging. Because the powdered metal heats slowly, longer-than-normal heating times are required to achieve uniform temperatures. This brings about some presintering, particularly in the powder nearest to the container wall. Once heated, the canned-powder forging blanks are press forged at speeds comparable with those used for aluminum forging. When the dies close, the high forging pressure is maintained for a brief dwell period to insure proper densification. The composite forgings are then cooled slowly to room temperature. The canning material is generally removed from the forged beryllium shape by either acid pickling or machining.

Cieslicki[3] described some of the procedures and problem areas in the production of a truncated conical shape by the canned-powder forging method. The container for such a part is made up of two concentric conical jackets with rings welded at the ends, as shown in the schematic drawing in Figure 22-3. During forging, the cap rings have a tendency to turn inward against the punch, causing the beryllium to flow toward the outer circumference and form a tapered fin. Because of differences in shrinkage

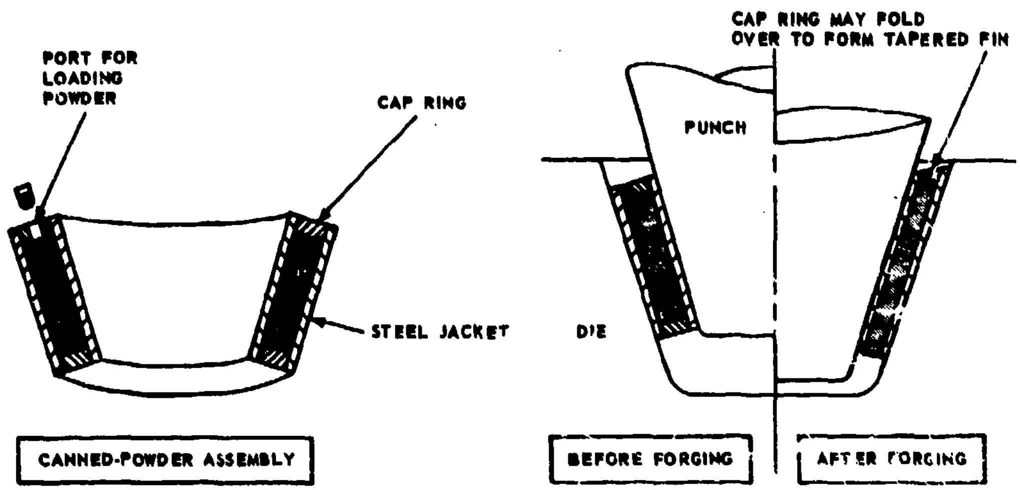


FIGURE 22-3. SCHEMATIC DIAGRAM OF CANNED-POWDER FORGING METHOD FOR CONICAL SHAPES[3]

between the steel and beryllium, the thinned-out beryllium section may crack during subsequent cooling. This can be a particular problem with cap rings of mild steel, which expands upon transformation. This problem can be minimized, however, by using Type 304 stainless steel for the rings.

Another problem when forging such hollow parts can arise from shrinkage of the beryllium cone onto the punch during pressing. This can occur particularly during the final pressing stage when the forged composite is thinnest and most readily chilled by the dies, and can set up sufficiently high stresses to cause cracking of the beryllium. Thus, it is important to maintain die temperatures as close as possible to the forging temperature.

Lubrication techniques for forging conical shapes can have a significant influence on the flow of both the canning material and the beryllium. The flow of both materials should be as uniform as possible from surface to surface. Thus, it is not uncommon to use different lubricants for the opposed dies to achieve balanced flow and minimize the chance for jacket failure.

Billet Forging

Vacuum-hot-pressed beryllium billets can be readily forged in closed dies, provided that a positive method for keeping the billet in compression is employed. The necessary restraint (to avoid cracking of the forgings from the high tensile stresses normally induced during forging) can be achieved in several ways

(1) Canning the billet in thick steel jackets

(2) Application of expendable steel support rings

(3) Carefully controlled die design and die staging.

The latter two methods employ bare billets which offer the advantages of lower billet preparation cost and greater definition of part geometry than for clad billets.

The steel-jacketing technique for forging hot-pressed billet is similar to the canned-powder forging method, as shown by Figure 22-4, which gives the details for forging the same beryllium cup shape described in Figure 22-2. It is important to shape the billet so that it will be subjected to high restraint throughout the forging cycle. As shown by the details in Figure 22-5 for forging the heat sink for the Project Mercury space capsule, this can be achieved by starting with a billet preformed to the same contour as the die.

Because of production and design limitations and the additional costs associated with the use of jackets, considerable effort has gone into the development of methods for forging unclad beryllium billet. The forging of a part by a combination of back extrusion and upsetting operations using hot steel support rings is illustrated in Figure 22-6. In this technique developed by Ladish Company[8], the steel rings confine the workpiece and provide restraint to the free surfaces during deformation.

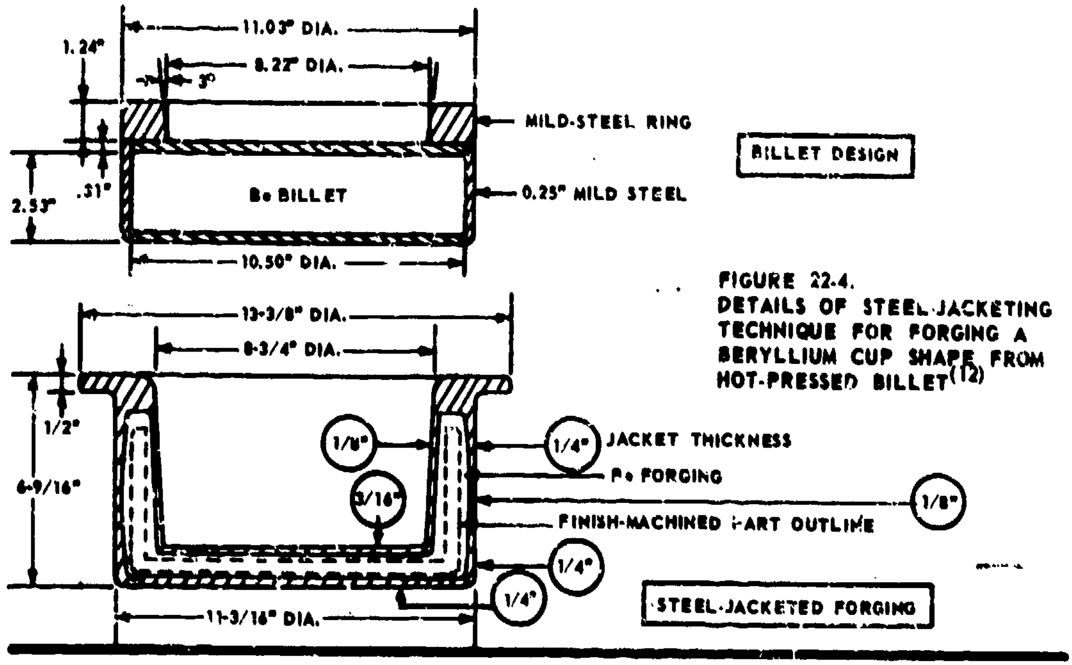


FIGURE 22.4.
DETAILS OF STEEL-JACKETING TECHNIQUE FOR FORGING A BERYLLIUM CUP SHAPE FROM HOT-PRESSED BILLET(12)

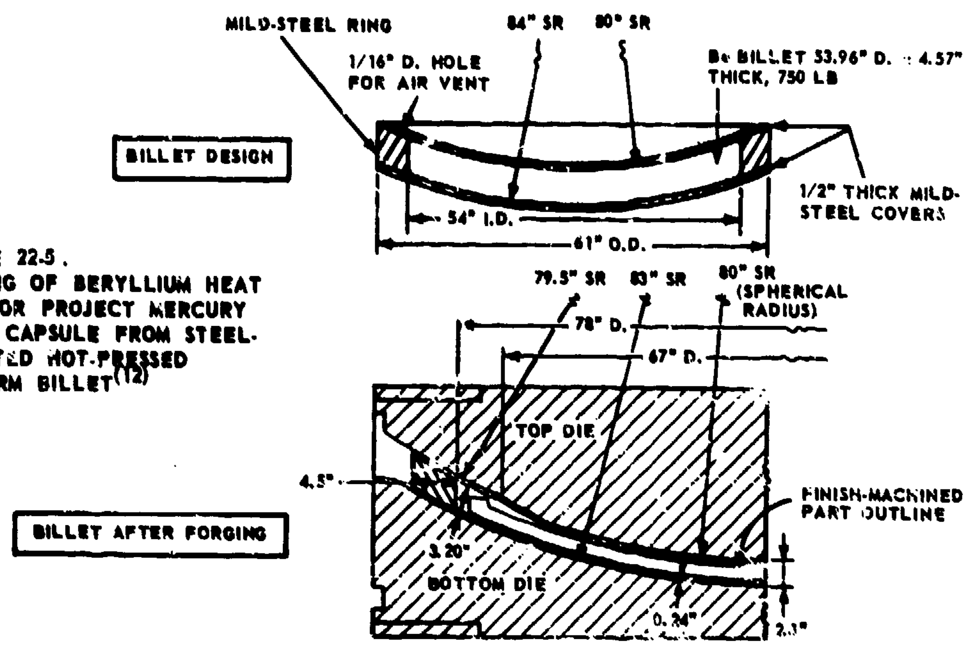


FIGURE 22.5.
FORGING OF BERYLLIUM HEAT SINK FOR PROJECT MERCURY SPACE CAPSULE FROM STEEL-JACKETED HOT-PRESSED PREFORM BILLET(12)

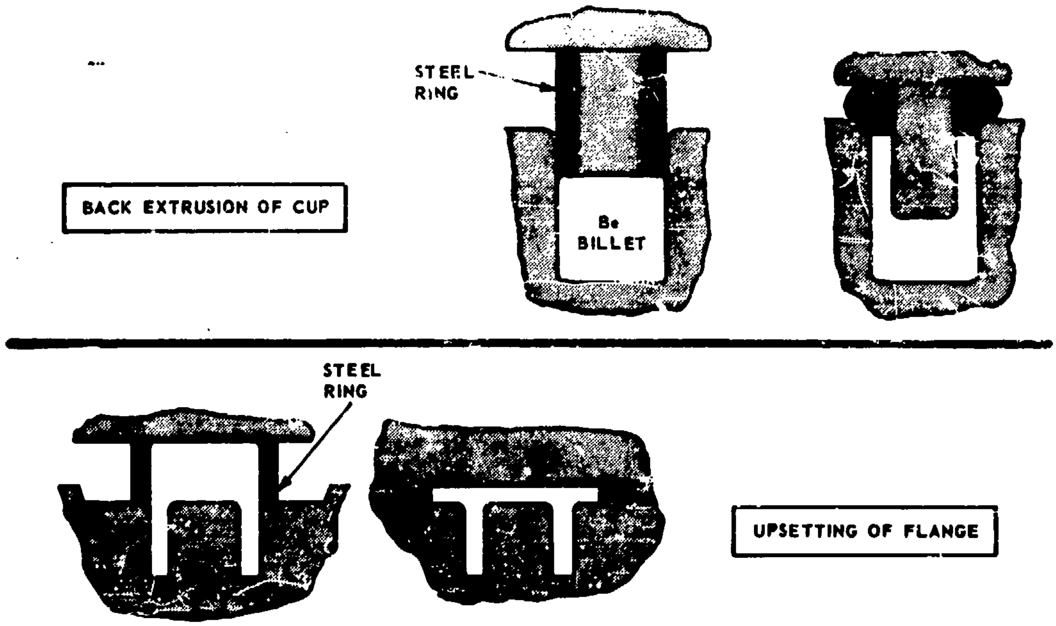


FIGURE 22-6. FORGING OF BERYLLIUM PART BY BACK EXTRUSION AND UPSETTING WITH STEEL SUPPORT RINGS(8)

With a maximum billet temperature of 1400 F (for reasons of forgeability, properties, and safety), the carbon-steel rings are heated to temperatures ranging from about 1600 F to 2000 F, depending on the restraint requirements. Dies are heated to 800 F and sprayed with a standard oil-base forging lubricant. The beryllium blanks are etched in a 3% H_2SO_4 - 3% H_3PO_4 acid solution and coated with a low-temperature glass-frit enamel. If two or more forging stages are required for producing a part, the blanks are vapor blasted, etched, and recoated between each operation.

Hot-pressed beryllium can undergo a limited amount of deformation without restraint. The amount of deformation can be increased in closed dies by progressively restricting or "choking" the die cavity so that the die wall offers restraint to the workpiece. The selection of die designs and lubricants is quite critical to avoid cracking that can result from a poor metal-flow condition or galling of the dies. Denny and McKeogh(13) demonstrated that a hub section with a circumferential web around the hub could be forged from a cylindrical hot-pressed block of beryllium. An important feature of the part design, shown in Figure 22-7, is the full-radius internal fillet. The gradual transition with this fillet design facilitates lubrication, reduces the amount of redundant shear deformation of the surface and the attendant shear stresses, and so aids metal flow. For small fillet radii, multiple forging stages would be required to gradually distribute the metal into the desired locations before the finish forge operation.

By exact control of the starting billet geometry and carefully designed die staging, Hornak and O'Rourke(10) showed that quite complex structural parts can be forged with reasonable success from hot-pressed beryllium without artificial restraint. An experimental aircraft bracket design of the configuration shown in Figure 22-8 was used in

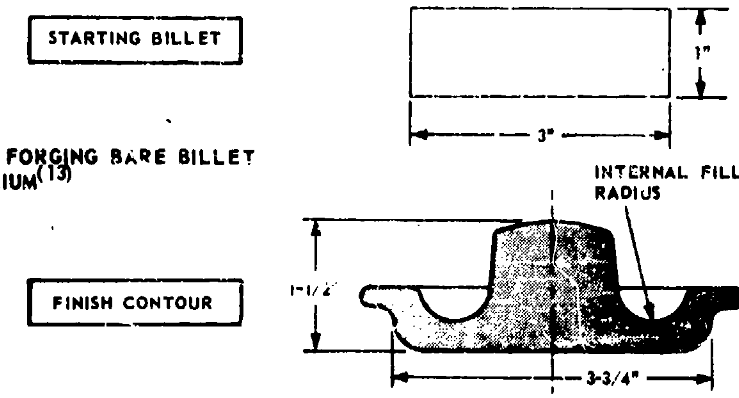


FIGURE 22-7.

SUCCESSFUL DESIGN FOR FORGING BARE BILLET OF HOT-PRESSED BERYLLIUM[13]

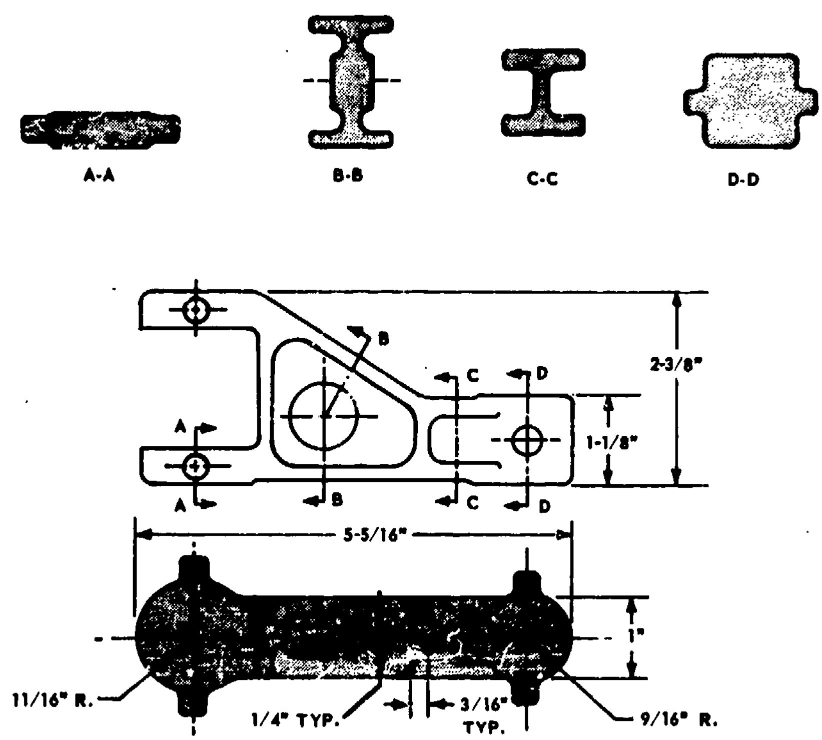


FIGURE 22-8. CONFIGURATION OF EXPERIMENTAL STRUCTURAL PART FORGED FROM UNCLAD HOT-PRESSED BERYLLIUM[10]

this study. The recommended fabrication procedure for this experimental part is summarized in Table 22-8, and the forging sequence is schematically illustrated in Figure 22-9.

Stress Relief. The term "stress relieving" involves heating to a suitable temperature, holding long enough to reduce residual stresses, and then cooling slowly enough to minimize the development of new residual stresses. For beryllium that has been fabricated in one manner or another, the deformation is accompanied by the accumulation of residual stresses of significant magnitude.[14] These stresses have important effects both on the strength of the structure under service loads and also on the propagation of cracking, once failure is initiated. Producers therefore recommend that all wrought shapes of beryllium be stress relieved following deformation. Temperatures in the range of 1300 to 1400 F are most effective in accomplishing adequate stress relief although stresses may be relieved with long exposures at temperatures as low as 1150 F. Temperatures above 1400 F are likely to cause recrystallization and grain growth, thus destroying the favorable properties of the wrought product. Usually stress-relieving times of 10 to 20 minutes are adequate.

TABLE 22-8. FABRICATION PROCEDURE FOR FORGING A STRUCTURAL PART FROM BARE HOT-PRESSED BERYLLIUM[10]

A. Forging Equipment

1. Hydraulic press
2. Hot-work tool steel dies at 800 F

B. Forging Stock

1. Vacuum-hot-pressed preform
2. Preform shape to be flat with perimeter to match blocker die cavity
3. Stock temperature - 1350 ± 25 F

C. Operations

1. Machine beryllium preforms
   (a) Visual inspection after light etch
   (b) Sonic and dye-penetrant inspection
   (c) Coat with lubricant

2. Block forge preforms with heavy flash
   (a) Remove adherent lubricant by vapor blast
   (b) Visual inspection after light etch
   (c) Sonic and dye-penetrant inspection
   (d) Condition surface by grinding out defects
   (e) Mill off heavy flash
   (f) Etch and coat with lubricant

3. Finish forge machined blocked parts without flash
   (a) Remove adherent lubricant by vapor blast
   (b) Visual inspection after light etch
   (c) Sonic and dye-penetrant inspection

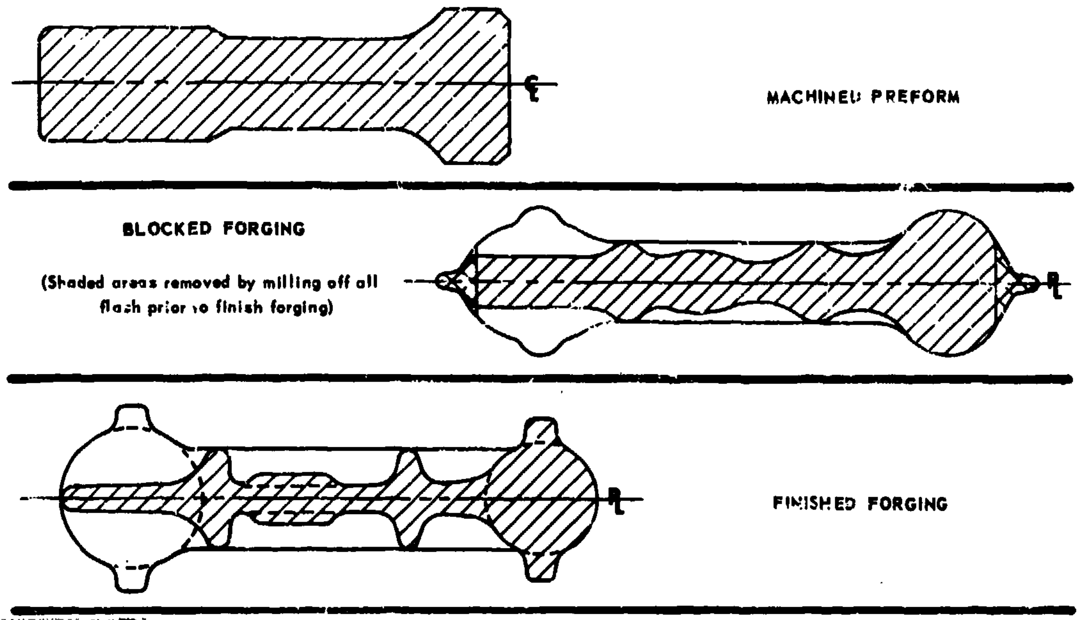


FIGURE 22-9. SCHEMATIC DIAGRAM OF FORGING SEQUENCE FOR EXPERIMENTAL BERYLLIUM STRUCTURAL PART[10]

Sometimes forgings are stress relieved at slightly higher temperatures and for longer times that approximate an annealing treatment. Hayes and Yoblin[15], for example, reported stress relieving forgings at 1400 to 1470 F for times ranging from 1/2 to 1 hour.

Safety. There are some basic requirements for the maintenance of health and safety in facilities handling beryllium. Good housekeeping and sensible shop-safety measures can offset the hazards associated with working with beryllium to a great extent. However, the absolute necessity of keeping beryllium dust out of the atmosphere is a subject that can never be neglected in handling this material.[[illegible], [illegible]7]

Practical general suggestions for working with beryllium might include the following:[18,19]

(1) Equipment and other surfaces on which dust may collect must be cleaned thoroughly and frequently. Use of smooth surfaces covered with glossy paint assists in such cleaning.

(2) Vacuum cleaning is the preferred procedure for removing beryllium dust from equipment and other plant areas. Central or movable vacuum equipment may be used, but air passing through such equipment must not be discharged into the work area.

(3) Dry sweeping and the use of compressed air for blowing dust from any surface must be avoided.

(4) Damp cloths must be used for wiping surfaces that cannot be adequately cleaned by vacuum.

(5) Streams of water should not be used in open work areas since they disperse contaminated mists or spray into air, particularly if working pressure is high.

For machining operations, every machine should be equipped with a special vacuum line near the tool for evacuating air at velocities usually ranging from 500 to 3000 feet per minute. Some machining operations such as milling, broaching, and rotary and bandsawing operations require the use of cutting fluids in addition to local exhaust ventilation to minimize dispersal of dusts. In such cases, provision must be made to collect the contaminated mists and sprays.

COMMENTARY ON FORGING DESIGN

Table 22-9 lists seven possible techniques for producing forgings and describes briefly their potential for producing various shapes and sizes of parts. The canned-powder forging technique probably offers the greatest size versatility, but it is limited to simple geometric shapes. Likewise, the technique of forging jacketed hot-pressed block is limited to simple shapes of varying sizes. Of these two techniques, the canned-powder method seems to be more useful for producing nonsymmetrical parts having bosses and other projections. This is because the hot-pressed preforms require complex, close-fitting jackets that are more difficult and costly to make.

The techniques for forging complex, hot-pressed, unjacketed preforms show a real potential for producing small forgings having a comparatively high degree of shape complexity. Perhaps preforms made by canned-powder forging technique will prove more forgeable or economical than hot-pressed preforms. Probably the most complex preforms could be made by slip casting powder slurries into absorbent molds, and then sintering the dried preforms to increase strength and density. A major problem with this type of preform would be the prevention of gas penetration when heating the preforms for forging. Approaches to this problem may involve plating, flame spraying, cold pressing, or other techniques for sealing the surface.

Drawing on successful practices for forging beryllium and experience with other low-forgeability materials, the over-all approach for designing beryllium forgings should include:

(1) Large fillet radii about two to four times that for aluminum forgings (blocker-type design)

(2) Generous corner radii - minimum of 1/8 inch and probably two times that for aluminum forgings

(3) Tapered webs, thickening toward outside at 3 to 5 degrees per surface

TABLE 22-9. SHAPE AND SIZE POTENTIAL FOR VARIOUS POSSIBLE METHODS OF FORGING BERYLLIUM

Based on an Industry Survey

Forging Method	Shape Potential	General Size Potential
1. Forging of canned powder	Generously contoured disks, hemispheres; hollow cylinders and cones; parts having small bosses and other short projections; simple structural shapes such as angles and channels	Section sizes up to 4 inches thick forged on a semiproduction basis, up to 12 inches thick on a developmental basis; disk shapes up to about 100-inch diameter
2. Forging of steel-jacketed hot-pressed block	Generously contoured disks, spherical surfaces, and shallow cones	Largest part produced is 80-inch-diameter spherical surface; size potential limited by capacity of vacuum hot-pressing facilities; section sizes up to about 12 inches thick
3. Forging of bare, hot-pressed block	Back extrusions, disks	Small parts with major dimension probably no larger than 5 or 6 inches; section thicknesses of less than 2 inches
4. Forging of bare, hot-pressed block using support rings	Back extrusions, disks hubs, cups, hemispheres	Section sizes of less than 2 inches; diameters up to 6 inches; back extrusions up to 10 inches long
5. Forging of bare, contoured, hot-pressed preforms	Low-profile disks (more refined detail than Method 3), small structural shapes with shallow profiles	Small parts ranging in size up to about 6-inch diameter; section sizes up to about 1 inch thick
6. Forging of slip-cast and sintered preforms	Intricate contours	Small parts; probably less than 3 pounds; section size limits probably on the order of 1 [illegible]
7. Reforging of bare preforms made by Method 1	Similar to Method 1, but more refined detail	Small parts, probably less than 10 pounds

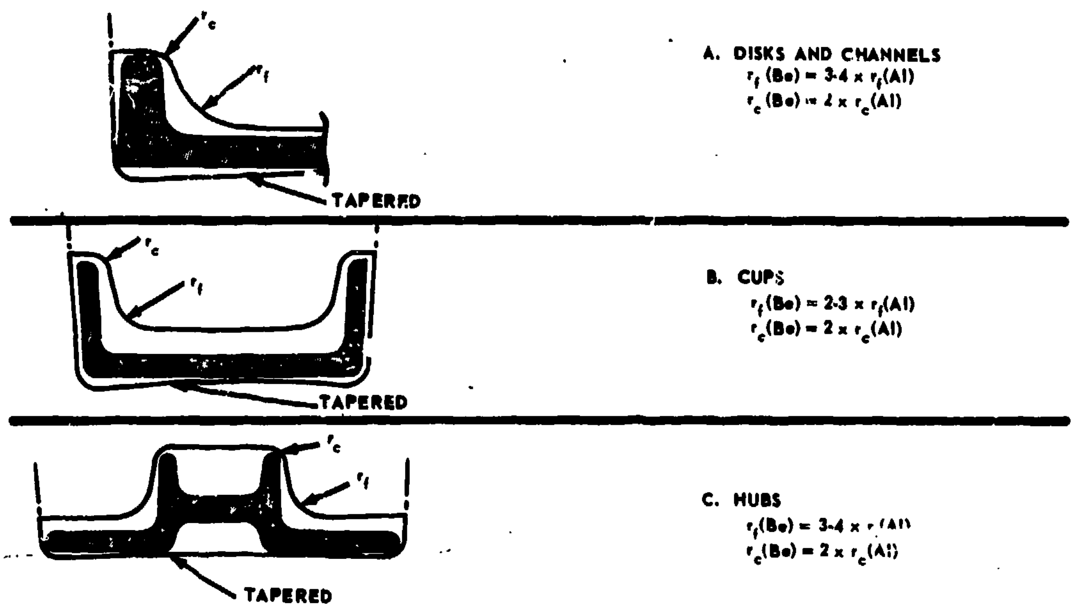
(4) Draft angles of 8° on inside surfaces and 5° on outside

(5) Section thickness of about 1/2 inch or larger

(6) Designs should avoid lateral flash; dies should be parted at the tops of ribs, rims, and other edge projections

(7) Bead at the parting line recommended if lateral flash unavoidable.

Figure 22-10 illustrates several typical sections found frequently in die-forging designs. The shaded designs in each view are those for aluminum-alloy forging.* The envelope outlines illustrate the designs recommended for beryllium. The generous fillet radii help provide for smooth metal flow. The tapered webs aid lubrication and promote lateral spreading of surface material. This ordinarily helps to maintain a greater degree of compressive stress at the unsupported edges during forging.



SHADED AREAS INDICATE ALUMINUM FORGING DESIGN

FIGURE 22-10. COMPARISON OF DESIGNS FOR ALUMINUM AND BERYLLIUM FORGINGS

*Refer to Chapters 3, 4, and 7 for discussions of aluminum forging design.

REFERENCES

(1) Wolff, A. K., Gelles, S. H., and Aronin, L. R., "Impurity Effects in Commercially Pure Beryllium", Paper No. 66, Conference on the Metallurgy of Beryllium, The Institute of Metals, London, England, 15 pp (October, 1961).

(2) Fenn, R. W., Jr., Glass, R. A., Needham, R. A., and Steinberg, M. A., "Beryllium-Aluminum Alloys", Paper presented at AIAA Fifth Annual Structures and Materials Conference, Palm Springs, California, pp 92-104 of Conference Proceedings (April 1-3, 1964).

(3) Cieslicki, M. E., "Beryllium Forging", Paper No. 2, Conference on the Metallurgy of Beryllium, The Institute of Metals, London, England (October, 1961).

(4) White, D. W., and Burke, J. E., The Metal Beryllium, American Society for Metals, Cleveland, Ohio (1955).

(5) O'Rourke, R. G., Hurd, J. M., Wikle, K. G., Beaver, W. W., "Mechanical Properties of Reactor Grade Beryllium at Elevated Temperatures", U. S. AEC, Physics Section, Report No. Coo-312 (August, 1956).

(6) Beaver, W., and Wikle, K. G., "Mechanical Properties of Beryllium Fabricated by Powder Metallurgy", AIME Transactions, 200 (1954).

(7) Kaufman, A. R., Gordon, P., Lillie, D. W., "The Metallurgy of Beryllium", ASM Transactions, 42 (1950).

(8) Hayes, A. F., and Yoblin, J. A., "Beryllium Forging Program", ASD Technical Report 62-7-647, Contract AF 33(600)-36795, Ladish Company (June, 1962).

(9) "Beryllium Forgings-Grade PF-20", Tentative Specifications, The Beryllium Corporation, Reading, Pennsylvania (August 1, 1962).

(10) Hornak, M. B., and O'Rourke, R. G., "Developing Technology to Forge Unclad Beryllium", Phase II Interim Report IR 7-973(II) on Contract AF 33(657)-8508, The Brush Beryllium Co., Cleveland, Ohio (November, 1963).

(11) Hodge, W., "Beryllium For Structural Applications", DMIC Report 168, Defense Metals Information Center, Battelle Memorial Institute (May 18, 1962).

(12) Hornak, M. B., and O'Rourke, R. G., "Developing Technology to Forge Unclad Beryllium", Phase I Interim Report on Contract AF 33(657)-8507, The Brush Beryllium Co., Cleveland, Ohio (September 10, 1962).

(13) Denny, J. P., and McKeogh, J. D., "Forging Unclad Beryllium", Journal of Metals, pp 432-433 (June, 1961).

(14) "Forming Beryllium", Bulletin 2140 published by The Beryllium Corporation, Reading, Pennsylvania.

(15) Hayes, A. F., and Yoblin, J. A., "New Concepts for Fabricating Beryllium Through Advanced Forging and Powder-Metallurgy Techniques", Paper No. 53, Conference on the Metallurgy of Beryllium, The Institute of Metals, London, England (October, 1961).

(16) "Workshop on Beryllium", Report of Seminar by the Kettering Laboratory, Department of Preventive Medicine and Industrial Health, College of Medicine, University of Cincinnati, Cincinnati, Ohio, Air Force Contract AF 33(600)-37211 (January, 1961).

(17) Breslin, A. J., and Harris, W. B., "Health Protection in Beryllium Facilities - Summary of Ten Years of Experience", Report No. HASL-36, U. S. Atomic Energy Commission, New York Operations Office (May 1, 1958).

(18) Hodge, W., "Some Notes on Safe Handling Practices for Beryllium", DMIC Memorandum 2, Defense Metals Information Center, Battelle Memorial Institute (September 22, 1958).

(19) Choluk, J., "Safe Way to Machine Beryllium", American Machinist/Metalworking Manufacturing, 107 (11), 82-84 (May 27, 1963).

APPENDIX A

FUNDAMENTALS OF PLASTIC DEFORMATION

APPENDIX A

FUNDAMENTALS OF PLASTIC DEFORMATION

Metals and alloys have played vital roles throughout the history of mankind, both as tools for fashioning other materials and as materials of construction. Ferrous alloys are the framework upon which our modern technological society is constructed. The use of metals has progressed from those found in nature in the metallic state (gold, silver, copper, iron-nickel meteorites), to the metals and alloys first produced by accidental smelting operations (during the bronze age, followed by the iron age), to metals easily smelted (lead, zinc, tin), and finally to metals won from their ores with difficulty (aluminum, magnesium, titanium, zirconium). Fabrication methods have advanced from simple forging and casting procedures to a multitude of primary and secondary forming operations, powder-metallurgy techniques, and considerably expanded casting techniques. The continuously increasing demands of designers on materials performance has led some writers to call the present era the Materials Age. Requisite to satisfying these demands is the ability to form metallic materials into the desired shape, which is based primarily on the phenomenon of plastic deformation.

A precise definition of a metal is not easy to formulate. The term "metal" refers to a class of more than 70 elements in the periodic chart, each with a characteristic set of properties. These characteristic properties arise from the nature of the bonding forces which hold crystals together; namely, interaction of an array of electrically charged atom cores (ions) with a free-electron gas. Some important manifestations of such "metallic bonding" are the following: (1) metallic luster, (2) opacity to light, (3) good thermal and electrical conductivity, (4) capacity for plastic deformation during relatively short times of loading, and (5) formation of a positive ion in the electrolysis of its compounds. A rigorous scientific criterion for a <u>pure</u> metal is that its temperature coefficient of electrical conductivity be negative, i.e., a metal becomes a poorer conductor as the temperature is raised. Insulators and semiconductors exhibit increasing conductivity with increasing temperature.

Aside from the factors of availability and economics, the two most important properties giving rise to the widespread use of metals and alloys are (1) their ability to be formed into useful shapes by plastic deformation and (2) their ability to develop a wide range of properties, particularly in terms of attractive combinations of strength and ductility, after suitable thermal and mechanical treatments. Of these two factors, the capacity for permanent changes in shape while maintaining coherency must be considered the most important. This is a particularly impressive property when one considers that metals are generally used as polycrystalline aggregates, each grain of which must accommodate to the deformation of its several immediate neighbors. Without this remarkable capacity for plastic deformation, forming of metals would be restricted to casting, powder-metallurgy techniques, and metal-removal processes (machining, grinding, electroforming). It is interesting to note that casting in a broader sense is a deformation process which takes advantage of the flowability of the liquid state (zero resistance to flow, or inability to maintain a shear stress) to fill a prepared cavity. Thus, the two most important forming processes for metals involve flow in the liquid state and flow in the solid state.

MECHANICAL RESPONSE OF PERFECT ELASTIC SOLID

All theoretical considerations of the flow and fracture of solids are based on the ideal elastic solid, one which is a perfect crystalline solid free of defects of any type. The ideal solid is crystalline by definition, meaning that a certain atomic arrangement is repeated indefinitely in space. The position of neighboring atoms is identical as we pass from atom to atom. Each atom for example, in the face-centered cubic or hexagonal-close-packed crystal structures has 12 close neighbors at characteristic distances. The processes of slip by shearing a layer of atoms rigidly over another, or fracture by separating two planes of atoms, requires the surmounting of the Coulombic forces of attraction and repulsion arising from the electrical fields associated with the atoms. It is useful to estimate the theoretical shear strength and fracture strength of an ideal elastic solid, which will represent the maximum attainable strength of solids.

Theoretical Shear Strength

The Frenkel(1) estimate of the theoretical shear strength, or resistance to flow by slip, is outlined below. The forces involved in translating a row of atoms over another to the next equilibrium position is illustrated in Figure A-1. As Row (1) moves over Row (2), a restoring force tending to pull Row (1) to its original position must be overcome. This restoring force reverses its sign as the midpoint position is passed and the row of atoms "snaps" into the next equilibrium position. The estimation of the theoretical shear strength depends on estimating the maximum restoring force.

The force-displacement relationship is approximated by a sine curve,

$$\tau = K \sin (2\pi x/b) \quad , \tag{A-1}$$

as illustrated in Figure A-2. The constant K is evaluated from the fact that the force-displacement curve must have a slope equal to that of shear modulus curve near the origin. Near the origin, it follows that

$$\tau = K\, 2\pi x/b \tag{A-2}$$

from the small-angle approximation.

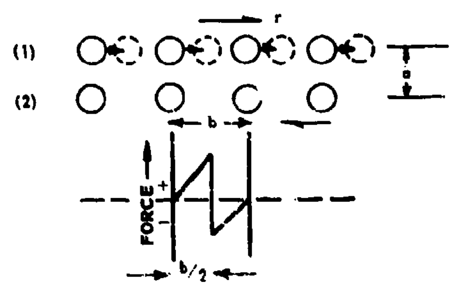


FIGURE A-1. SCHEMATIC ILLUSTRATION OF FORCES IN THE TRANSLATION OF ONE ROW OF ATOMS OVER ANOTHER

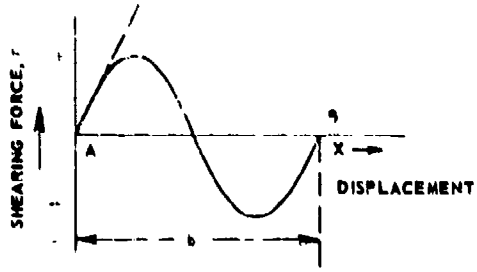


FIGURE A-2. APPROXIMATION OF FORCE-DISPLACEMENT CURVE BY SINE CURVE

From elasticity theory,

$$\tau = \mu \cdot x/a, \qquad \text{(A-3)}$$

where

μ = shear modulus

x/a = shear strain.

Thus

$$K\, 2\pi x/b = \mu \cdot x/a \qquad \text{(A-4)}$$

or

$$K = \frac{\mu b}{2\pi a} \quad . \qquad \text{(A-5)}$$

Then

$$\tau = \frac{b}{a}\frac{\mu}{2\pi} \cdot \frac{2\pi x}{b} \quad .$$

The critical shear stress comes at $x = b/4$, yielding

$$\tau_m = \frac{b\mu}{a2\pi} \quad . \qquad \text{(A-6)}$$

Since $a \approx b$, the theoretical shear strength is about $\mu/2\pi$, or one-sixth of the shear modulus. Slip occurs in pure metals at very small shear stresses, about $10^{-4}\,\mu$, which is about a thousandfold less than the theoretical shear strength.

Theoretical Fracture Strength

Fracture in the ideal elastic solid requires that the atomic forces of interaction be overcome. Consider a line of atoms in such a perfectly crystalline solid as illustrated schematically in Figure A-3.

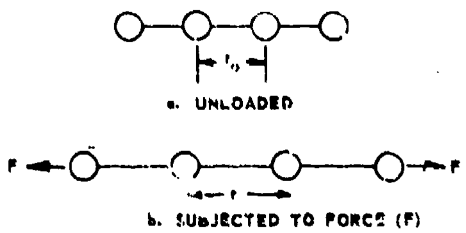


FIGURE A-3. LINE OF ATOMS IN A PERFECT CRYSTALLINE SOLID

The electrical fields associated with each atom give rise to potential fields of attraction and repulsion which are functions of the distance of separation of the atoms. The potential energy[2] of interaction, $\phi(r)$, can be expressed as:

$$\phi(r) = -\frac{\alpha}{r^m} + \frac{b}{r^n} \quad , \qquad \text{(A-7)}$$

where r is the distance of separation, α and m are constants for the attraction potential, and b and n are constants for the repulsion potential. It should be noted that $n < m$, since the repulsion potential (b/r^n) must increase more rapidly with increasing r in order that an equilibrium can be established between the atoms at r_0.

Equation (7) may be differentiated with respect to the distance of separation (r) to obtain the interacting forces:

$$F(r) = \frac{d}{dr}\phi(r) = -\underbrace{\frac{\alpha}{r^{(m+1)}}}_{\text{(attraction)}} + \underbrace{\frac{b}{r^{(n+1)}}}_{\text{(repulsion)}} \quad . \qquad \text{(A-8)}$$

This relationship is plotted in Figure A-4.

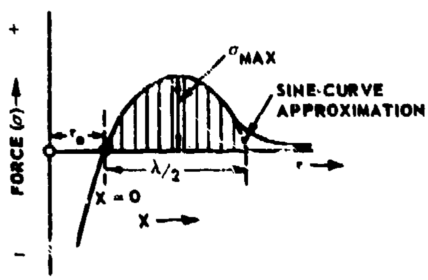


FIGURE A-4. RESULTANT-FORCE CURVE BETWEEN TWO ATOMS AND APPROXIMATION OF CURVE FROM $r = r_0$ BY A SINE CURVE

The sine-curve approximation is given by

$$\sigma = \sigma_{max} \sin (2\pi x/a) \quad . \qquad \text{(A-9)}$$

The calculation of the theoretical strength of an ideal elastic solid is based on the hypothesis that all the energy of separation is available for the creation of two new surfaces, which would amount to the surface energy of the new surfaces. The total energy available for the fracture process corresponds to the area under the curve from the equilibrium distance of separation ($r = r_0$) to infinite separation ($r = \infty$). The calculation so carried out[3] yields the result

$$\sigma_{max} = \sqrt{\frac{ES}{a}} \quad , \qquad \text{(A-10)}$$

where

σ_{max} = theoretical fracture strength

E = modulus of elasticity

S = surface energy per unit area

a = lattice parameter, or the equilibrium spacing r_0.

The theoretical fracture strength of perfect crystalline solids, accordingly, is directly proportional to the square root of the modulus of elasticity and the surface energy, and inversely proportional to the square root of the lattice parameter of the crystal. Substituting values of these three properties, it is found that the theoretical fracture strength would be expected to be about one-tenth of the modulus of elasticity. The expected values for iron, copper, and aluminum would be expected to be 3×10^6 psi, 1.6×10^6 psi, and 1.0×10^6 psi, respectively. In real metals, the theoretical elastic response terminates at a stress of 0.1 to 0.001 σ_{max} by the onset of plastic flow, which terminates by fracture after varying amounts of plastic flow. It is assumed here that plastic flow is a prerequisite for the onset of fracture in real metals and solids.

Both the resistance to plastic flow by slip and the resistance to fracture in real metals thus are several orders of magnitude lower than the values characteristic of the ideal elastic solid. In the following discussion, the fundamental problem we will be concerned with in understanding flow and fracture in real metals and alloys is to rationalize the discrepancy between theoretical and observed values. This rationalization will be based on the hypothesis that real metals contain discontinuities in an elastic continuum which are regarded as viscous elements in an elastic matrix. These elements can relax under load to produce plastic deformation according to a limited number of basic mechanisms. The nature of these relaxation centers will be examined.

THE RHEOLOGICAL MODELS FOR REAL SOLIDS

In the field of rheology, which deals with general laws of flow of materials, real solids are rationalized as combinations of two basic elements, namely, the perfectly elastic or Hookean solid, and the perfectly viscous, or Newtonian fluid.[4] The Hookean solid is one which has the following relationship between stress (σ) and strain (ϵ),

$$\sigma = K\epsilon \quad , \qquad \text{(A-11)}$$

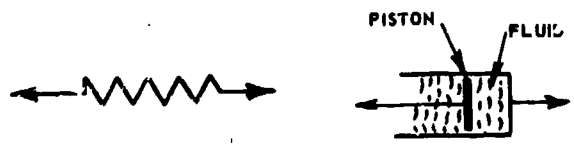
where K is a constant of the material. A linear relationship between stress and strain is assumed and an instantaneous response of strain to stress is inherent. This element is represented by a spring, as shown in Figure A-5a. The Newtonian fluid follows the relationship

$$\sigma = \eta\dot{\epsilon} \qquad \text{(A-12)}$$

where η is a material constant and $\dot{\epsilon}$ (or ϵ/t) is the strain rate. The Newtonian fluid is represented by a dashpot, as shown in Figure A-5b, and is the basis for relaxation under an imposed load, since if $\epsilon = 0$ when $t = 0$,

$$\epsilon = \sigma t/\eta \quad . \qquad \text{(A-13)}$$

Rheological models of the response of real solids to stress are built up as combinations of springs and dashpots.



a. THE HOOKEAN SOLID

b. THE NEWTONIAN FLUID

FIGURE A-5. THE BASIC ELEMENTS IN RHEOLOGICAL MODELS

There are two important deviations from the elasticity of the ideal elastic solid, namely <u>anelasticity</u> and <u>inelasticity</u>. The state of anelasticity is one in which the stresses are not sufficiently high to induce gross plastic flow and is a condition in which strain is not uniquely a function of stress, as is the case with an ideal elastic solid. Such a condition leads to a time dependency of strain, giving rise to internal-friction effects and strain-rate sensitivity in real solids. Figure A-6 is an example of a model that illustrates anelastic effects in the "general linear substance" An instantaneous elastic strain is experienced on application of a load. The remaining increment of elastic strain is gained as the dashpot relaxes with time. In the absence of the dashpot, the full strain expected by elastic theory would be developed instantaneously. The anelasticity of a body gives rise to a hysteresis effect when cycling in the elastic range, as illustrated in Figure A-7. The anelastic properties result in a mechanical hysteresis in stress cycling of such a body.

Inelasticity is the property of a substance which enables it to undergo permanent, irreversible strain as a result of imposed loads. In more common terms, the substance can undergo plastic deformation. In the spring and dashpot models, at least one dashpot, or viscous element, must be connected in series with the other elements. Figure A-8 shows two common models that illustrate the inelasticity of the Maxwell substance and

the Bingham substance. The Maxwell substance when loaded responds instantaneously in the elastic element, and the inelastic strain increases as a linear function of elapsed time. The Bingham substance differs from the Maxwell substance in that a frictional element simulates a yield strength; the response is elastic until the frictional resistance is overcome.

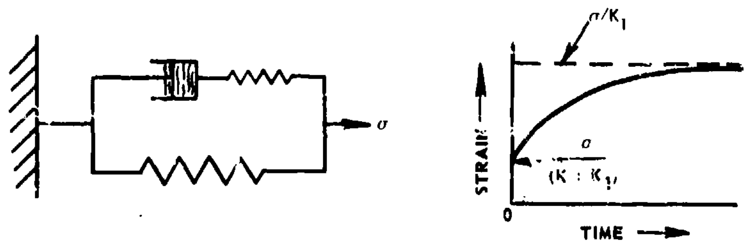


FIGURE A-6. CHARACTERISTICS OF THE GENERAL LINEAR SUBSTANCE UNDER THE INFLUENCE OF A CONSTANT STRESS, σ

K AND K_1 ARE MATERIAL CONSTANTS.

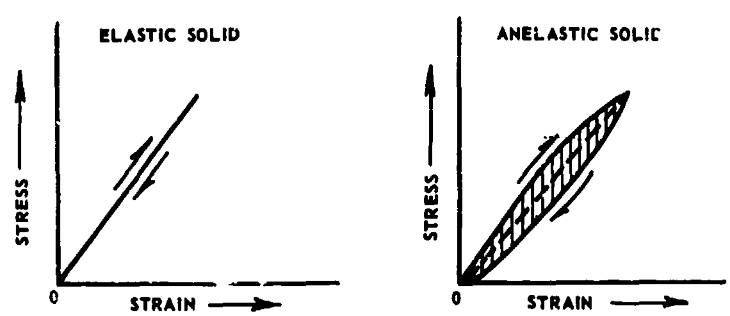
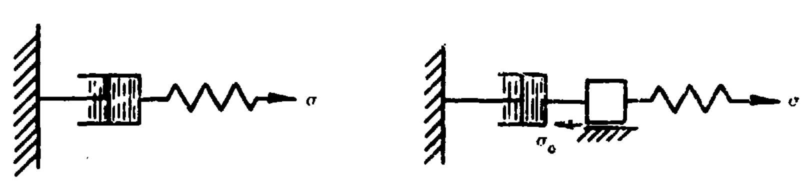


FIGURE A-7. CYCLING OF ELASTIC AND ANELASTIC BODIES

No energy is dissipated in cycling an ideal elastic body, whereas energy proportional to the enclosed area is dissipated per cycle in the anelastic body.



a. THE MAXWELL SUBSTANCE   b. THE BINGHAM SUBSTANCE

FIGURE A-8. RHEOLOGICAL MODELS POSSESSING THE CHARACTERISTIC OF INELASTICITY

The above are but a few of the models composed of springs and dashpots that have been formulated to simulate the response of real materials to imposed loads. It is important to note that, in such an approach, no physical characterization is made of the viscous elements; all are characterized by dashpots. Real metals and alloys possess several recognized types of viscous elements interspersed in an elastic matrix. The more important types of viscous elements, or relaxation centers, in real metals are reviewed in the next section.

DEFECT STRUCTURE IN REAL METALS

The viscous elements in real metals and alloys are classified as crystal defects. A defect is any deviation from a perfectly regular lattice or structure. The localized deviations from perfect periodicity result in loss of the characteristic high degree of order and produce disordered liquidlike structures that are viscous in nature and can relax under imposed loads. The crystal defects are classified as (1) point defects, (2) line defects, and (3) surface defects.

Point Defects

The simplest type of defect is one in which an atom is removed from a normal lattice site, leaving a vacancy. The two types of vacancy defects are illustrated in Figure A-9. The Schottky defect is a lattice vacancy with the missing atom displaced to the free surface. This lattice vacancy, dissociated from the displaced atom, is the type normally present in metal crystals. The Frenkel defect is a lattice vacancy and an interstitial atom pair, i.e., the atom which vacated a normal lattice site is located in some nearby interstitial space. Such defects are characteristic of metals that have suffered radiation damage. Such a structure is highly strained (disordered) around the interstitial atom because the atom fits very poorly in the available interstitial space.

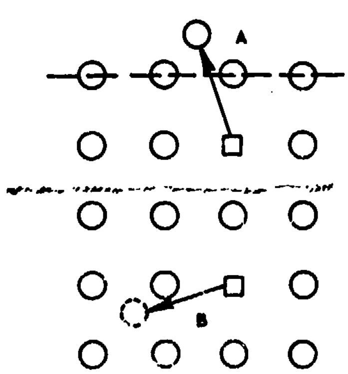
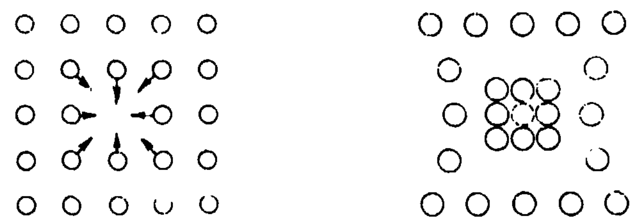


FIGURE A-9. BASIC LATTICE VACANCY DEFECTS IN CRYSTALS

A represents the Schottky defect and B represents the Frenkel defect

The Schottky defect is normally of more interest in metals than the Frenkel defect. When an atom is removed from the lattice, the local equilibrium of forces is disturbed and the neighboring atoms "collapse" on the hole to re-establish equilibrium of forces, as illustrated in Figure A-10. This collapsed structure around a vacancy creates an element of liquid structure according to the "hole" theory of liquids and was termed a "relaxion" by Nachtrieb. The disorder around the vacancy gives it a higher inherent mobility as compared with that around a hole with an unrelaxed structure.



a. AN ATOM IS REMOVED FROM A LATTICE SITE

b. ATOM DISPLACEMENTS TO RE-ESTABLISH EQUILIBRIUM OF FORCES

FIGURE A-10. ATOMIC ADJUSTMENT AROUND A LATTICE VACANCY (SCHOTTKY DEFECT) CREATING LOCALIZED DISORDER

That lattice vacancies are normally expected to be present in a crystal follows logically. As holes are mixed into a crystal, entropy increases as a result of the entropy of mixing, which in turn lowers the free energy of the crystal. An equilibrium concentration of vacancies (n) is expected at each temperature, which can be calculated by writing the free-energy change in introducing M vacancies in N lattice sites (energy required to remove an atom to the surface minus the temperature times the entropy of mixing), taking the derivative of this free-energy change with respect to M, equating the derivative to zero, and solving for n. The form of the expression so obtained for the equilibrium number of vacancies is

$$n \approx Ne^{-\frac{E_s}{kT}} , \qquad \text{(A-14)}$$

where E_s is the energy to remove an atom to the surface, k is Boltzmann's constant, and T is the absolute temperature. It is clear that the higher the temperature, the larger is the equilibrium concentration of vacancies.

Vacancies in metal lattices are important in several respects. The exchange of a vacancy with a neighboring atom constitutes the basic process of self-diffusion in pure metals and in substitutional alloy diffusion. Such an interchange also constitutes the process of the important mechanism of "climb" in the creep of metals (time-dependent plastic flow) at elevated temperatures. It is now recognized that clusters of two, three, or more vacancies are very mobile in metal crystals. Large groups of vacancies can collapse into dislocation rings; this movement constitutes slip in metals. It is generally recognized that the process of plastic flow produces a large number of vacancies which can coalesce into pores that join up through the process of plastic flow to constitute the mechanism of ductile fracture in metals.

Line Defects

Lattice disregistry which is manifested by a straight or curved line in space is termed a dislocation line. These dislocation lines are formed by two basic types of dislocations; namely, the edge dislocation and the screw dislocation, illustrated in Figure A-11. The edge dislocation is formed in a lattice by introducing an extra row of atoms. The screw dislocation is a spiral-ramp configuration along a particular direction called the line of dislocation. Schematic atomic models of these dislocations are presented in Figure A-12. Dislocations in crystals occur as lines terminating at free surfaces or closed loops in the crystal and are generally composed of combinations of edge and screw components. The movement of such lines or loops on certain crystallographic planes is believed to constitute the basic mechanism of slip. It is clear from Figure A-12a that, when an edge dislocation reaches a free surface, the upper part will project one atomic distance over the lower part. This constitutes a unit slip. The dislocation line is particularly mobile, since the restoring forces essentially cancel out because of the symmetry of the configuration.

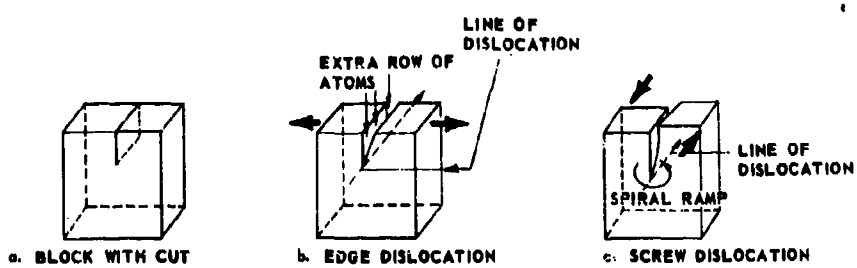
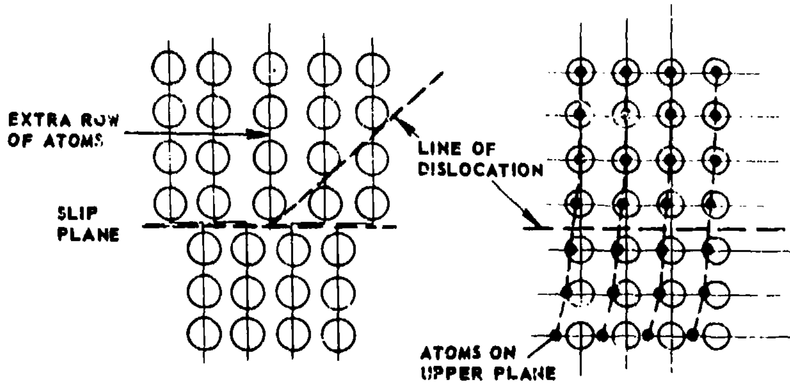


FIGURE A-11. SCHEMATIC DEPICTION OF EDGE AND SCREW DISLOCATIONS

Surface Defects

Neglecting the external surfaces, the concept of surface defects includes all types of interfaces in the interior of the polycrystalline aggregate that constitutes the usual metal or alloy. The usual states of subdivision are illustrated in Figure A-13. Three types of interfaces are depicted; namely, the grain boundaries, the subgrain boundaries, and the mosaic block boundaries. The energy of such boundaries is dependent on the degree of misorientation of the adjacent units. Grain orientations differ by several degrees of arc and are called high-angle boundaries (high energy), whereas subboundaries differ by degrees and minutes of arc, and mosaic blocks (about 10^{-4} cm on a side differ by minutes and seconds of arc. The latter two are called low-angle boundaries (low energy) as depicted in Figure A-14. These boundaries are transition zones from one crystal orientation to another orientation and, accordingly, possess disordered liquidlike structure and are rightfully considered viscous elements. Special interfaces are those formed with the matrix, by twins and stacking faults, and will be discussed in a later section.



a. EDGE DISLOCATION

b. SCREW DISLOCATION

FIGURE A-12. ATOMIC MODELS OF THE EDGE AND SCREW DISLOCATIONS

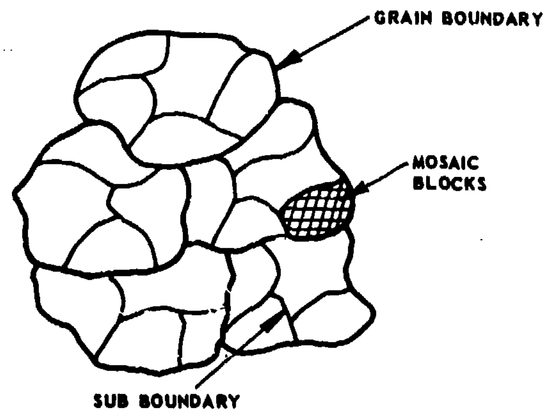


FIGURE A-13. SUB-DIVISION OF STRUCTURE OF A METAL
POLYCRYSTALLINE AGGREGATE

GRAINS RANGE IN DIAMETER FROM 20 TO 250 MICRONS.

Thus, all three basic classes of crystal defects can be regarded as viscous elements, dispersed in an elastic matrix, which are potential relaxation centers that can be activated under load. The single or couple relaxation of these elements gives rise to the observed anelastic and plastic response of real metals when subjected to loading. This deformation finally becomes unstable and is terminated by fracture occurring by one of several mechanisms.

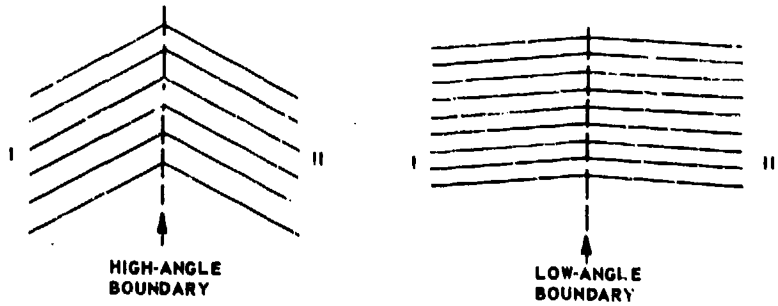


FIGURE A-14. ILLUSTRATION OF HIGH-ANGLE AND LOW-ANGLE BOUNDARIES

ELEMENTS OF DISLOCATION THEORY

The atomistic aspects of plastic deformation in solids currently are considered largely in terms of the dislocation theory. The concept of the slip process of plastic flow as the movement of dislocations serves to reconcile the difference between the theoretical and the observed strengths of solids and has had reasonable success in rationalizing the flow processes and some aspects of fracture initiation. Only the rudiments of the dislocation theory will be reviewed here. There are several excellent books available on the subject.(5)

The concept of edge and screw dislocations was introduced in the previous section dealing with the basic types of defects found in real solids, and are illustrated in Figure A-12. A basic idea in the theory is that of dislocation lines and loops. These lines and loops are composed of combinations of edge and screw dislocations; the lines may be straight or curved in space. It is a geometric requirement that the dislocation line cannot end within a crystal or a grain in a polycrystalline aggregate, but must terminate at a free surface or grain boundary, or form a closed loop (Figure A-15). It is also a geometric requirement that the termination of an edge dislocation generates a screw dislocation and, conversely, that the termination of a screw dislocation generates an edge dislocation, as illustrated in Figure A-16. The components of a dislocation line can be correctly thought of as the dislocation in the edge of the screw orientation. Both edge and screw segments can be of two types: edge segments can be positive or negative, depending on whether the extra row of atoms is above or below the slip plane, and screw segments can be clockwise or counter-clockwise. The dislocations interact in the usual manner: attraction between unlike signs and repulsion between like signs.

The dislocation line or loop is the boundary line separating the slipped from the unslipped portion of the slip plane, as illustrated in Figure A-17. The basic characteristic of the dislocation line is the Burgers vector ($\bar{b}$), which is the vector describing the distance and direction of the translation in a unit slip process of the plane of atoms above the slip plane across the plane below the slip plane. The movement of an edge dislocation is restricted to the plane and direction in this plane that gives the smallest Burgers vector. Slip planes and slip directions can be rationalized in this manner. The movement of a pure screw dislocation has no particular geometric limitations. The Burgers

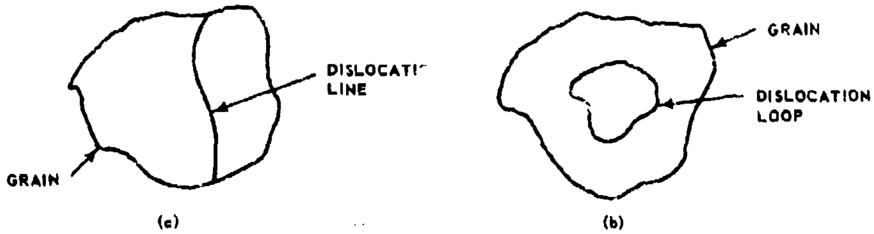


FIGURE A-15. ACCEPTABLE GEOMETRIC CONFIGURATIONS OF DISLOCATION LINES IN CRYSTALS

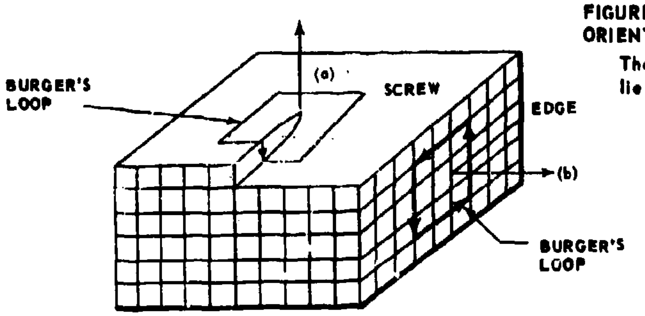


FIGURE A-16. CHANGE OF SCREW ORIENTATION TO EDGE DISLOCATION. The directions of the two dislocations lie at right angles (after W. J. Bollman).

(a) DIRECTION OF SCREW

(b) DIRECTION OF EDGE

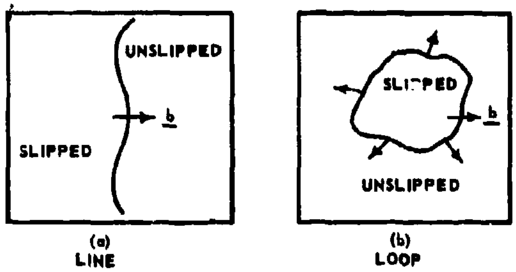


FIGURE A-17. THE DISLOCATION LINE AS THE BOUNDARY BETWEEN THE SLIPPED AND UNSLIPPED PORTION OF THE SLIP PLANE

vector is the same at every point on a dislocation line or loop. The force, per unit length of dislocation, is given by

$$F = \tau b \quad , \qquad \text{(A-15)}$$

where

τ is the shear stress on the slip plane, and

b is the strength of the dislocation.

That only a small force is necessary to move a dislocation can be understood from examination of the atomic arrangement of an edge dislocation. It is evident that corresponding atoms on each side of the extra row of atoms have restoring forces of opposite

sign, essentially cancelling each other out. This allows the atomic disregistry to pass through the crystal at low stresses. The Burgers vector of dislocations determines the energy of these defects and also the strength of interactions among them.

A simple viewpoint on the hardening of metals by plastic straining (strain hardening) was introduced by G. I. Taylor(6) in terms of a concept of a super-lattice of dislocations having like and unlike signs. The greater the density of dislocations, the stronger the interaction between them and the greater the resistance to plastic flow. The strain hardening of metals thus requires the generation of new dislocations, and the softening of a metal requires the reduction of the dislocation density.

New dislocations are generated by surface sources, internal sources such as inclusions, and the Frank-Read source, which is a specific type of internal source. The Frank-Read source pictured in Figure A-18 will be described, since it is the most widely recognized type of source. Segment A-B represents a pinned segment of a dislocation subjected to force F. The force causes the segment to bow out to the last stable position (2), at which point it loops around the pinning Points A and B and joins in position (5) at Point C. The loop moves off and the remaining segment returns to position A-B. Each source can generate many loops on being activated. The loops may pile up at an obstacle and exert a back stress which eventually stops the operation of the source, as shown in Figure A-19. This concept differs from the simple one suggested by G. I. Taylor.

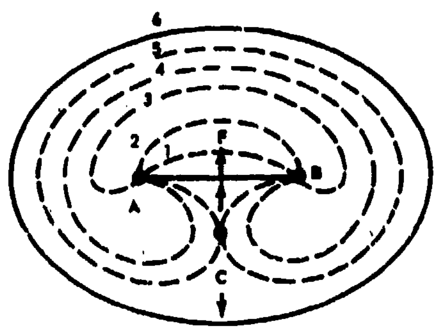


FIGURE A-18. OPERATION OF A FRANK-READ SOURCE, WHICH PRODUCES DISLOCATION LOOPS

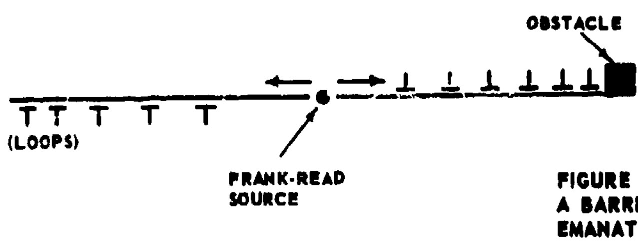


FIGURE A-19. CONCEPT OF PILE-UP AT A BARRIER OF DISLOCATION LOOPS EMANATING FROM A FRANK-READ SOURCE

In summary, the hardening of metals by plastic straining requires an increase in the density of dislocations, and the softening requires a decrease in the dislocation density. Reduction of dislocation density can occur by annihilation of unlike-sign dislocation pairs, by dissipation at a free surface, and presumably by disappearance back into

the source. The theoretical strength is expected if the dislocation density goes to zero the observed range of dislocation densities is from 10^6 to 10^{12} lines per square centimeter.

Dislocation theory has been used to rationalize many flow and fracture phenomena in metals including the following: slip, twinning, kinking, deformation bands, formation of stacking faults, cross slip, interface structures, strain hardening, polygonization, recovery, strain aging, the yield point, delay time, strain-rate sensitivity, embrittlement of body-centered cubic metals with reduction of temperature, creep, fracture initiation, and fatigue initiation.

DEFORMATION PROCESSES IN METALS

Plastic deformation occurs in metals by several mechanisms. It is convenient to discuss deformation in terms of two temperature ranges: (a) temperatures below Tm/2, where Tm is the melting point in degrees Kelvin, and (b) temperatures above Tm/2. Range (a) is usually referred to as the low-temperature range, and Range (b) as the high-temperature range. This temperature division has a physical basis, since self-diffusion rates become appreciable at Tm/2, which corresponds roughly with the temperature above which cold-worked metals are softened by recovery and recrystallization. Plastic deformation in this high-temperature range is referred to as "hot working" since softening keeps up with strain hardening and no residual hardening persists. Another important distinction is that metals and alloys deform continuously with elapsed time in the high-temperature range at nominal loads, a process referred to as "creep". The high-temperature range, therefore, is also referred to as the "creep range".

Deformation Mechanisms in the Low-Temperature Range

The deformation mechanisms in the low-temperature range (T < Tm/2) include slip, twinning, deformation bands, and kinking. Slip and twinning are the most important modes of deformation and account largely for the plastic strain that is generated before fracture. Kinking is of minor importance in the deformation of polycrystalline metals and will not be discussed here.

Slip. Deformation by slip can be said to take place by the movement of lamellae of the crystals over each other, much like the shearing of a deck of playing cards. The elements of the slip process for a single crystal are illustrated in Figure A-20. The slip

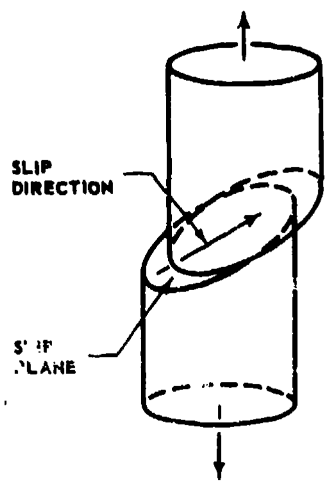


FIGURE A-20. SLIP SYSTEM FOR A SINGLE CRYSTAL

translation takes place on a definite slip plane in a definite crystallographic direction, the combination of these two elements constituting a slip system. The slip systems operative in the common crystal systems in metals are listed in Table A-1. It is to be noted that, although several slip planes may be available in a given crystal structure, the slip direction remains the same. New slip systems may be introduced at elevated temperatures, but the slip direction remains invariant. The slip planes for the three most common crystal structures are illustrated in Figure A-21.

TABLE A-1. SLIP SYSTEMS IN METALS[7]

Structure	Metal	Low Temperatures		High Temperatures	
		Plane	Direction	Plane	Direction
Face-centered cubic	Al	(111)	[101]	(100)	[101]
	Cu	"	"		
	Ag	"	"		
	Au	"	"		
	Ni	"	"		
	Cu-Au	"	"		
	αCu-Zn	"	"		
	αCu-Al	"	"		
	Al-Cu	"	"		
	Al-Zn	"	"		
	Au-Ag	"	"		
Body-centered cubic	α-Fe	(110)	[111]		
		(112)	"		
		(123)	"		
	Mo	(112)	"	(110)	[111]
	W	(112)	"		
	K	(123)	"	(123)	[111]
	Na	(112)	"	(110)	[111]
				(123)	[111]
	β Brass	(110)	"		
	α Fe-Si; 5% Si	(110)			
Close-packed hexagonal	Mg	(0001)	$[2\bar{1}\bar{1}0]$	(0001)	$[2\bar{1}\bar{1}0]$
				$(10\bar{1}1)$	$[2\bar{1}\bar{1}0]$
	Cd	"	"		
	Zn	"	"		
	Be	"	"		
	Zn-Cd	"	"		
	Zn-Sn	"	"		
Rhomobohedral	Bi	(111)	$[10\bar{1}]$		
	Sb				
	Te	(1010)			
	Hg	(100) and Complex			
Tetragonal	β-Sn	(110)	001	(110)	[111]
	(White)	(100)	001		
		(100)			
		$(10\bar{1})$	101		
		(121)	$10\bar{1}$		

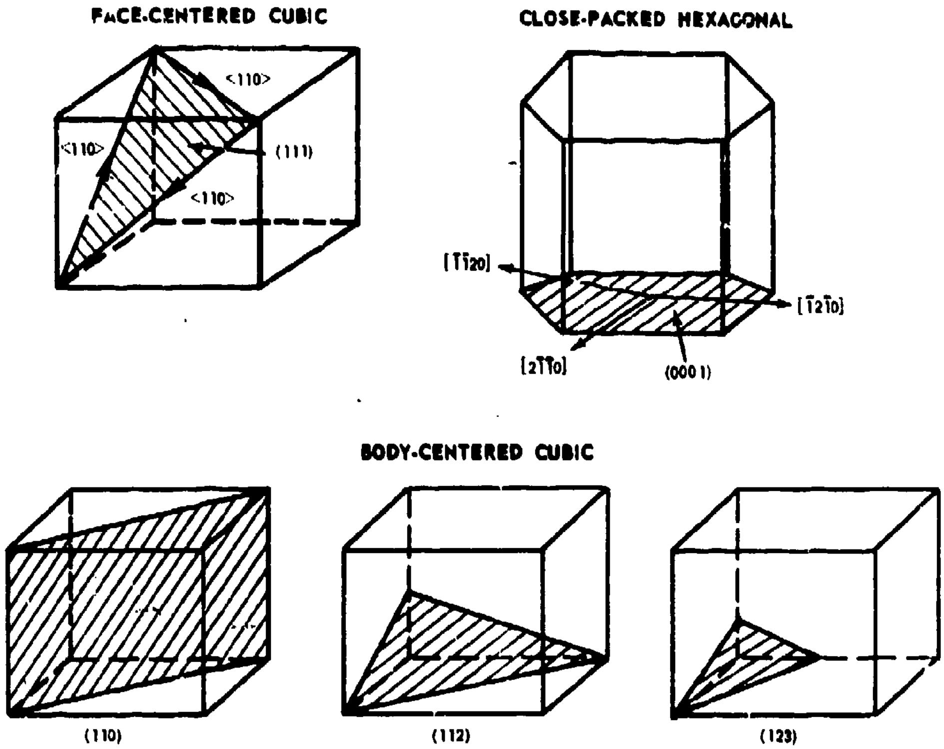


FIGURE A-21. SLIP PLANES IN THE COMMON CRYSTAL STRUCTURES IN METALS

The translation associated with slip causes the portion of the crystal above the slip plane to be displaced rigidly across the lower portion, as shown in Figure A-22. A unit slip gives rise to a displacement of one interatomic distance. In terms of dislocation theory, this displacement results when a dislocation line or loop reaches a free surface. Multiple displacements on a slip plane produce surface displacements, which enables slip lines to be seen with an optical microscope on previously polished surfaces. On repolishing of the surface, the discontinuity is removed and the slip lines no longer can be observed. Slip-line traces can be straight or wavy, depending on the crystal system.

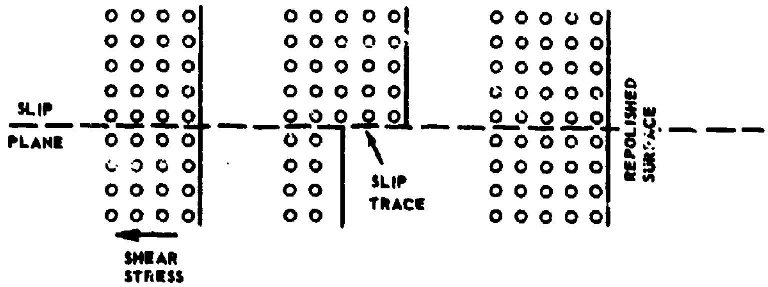


FIGURE A-22. DISPLACEMENT OF LATTICE LAMELLAE IN THE SLIP PROCESS

Twinning. The process of twinning results in a limited amount of plastic deformation, but is particularly important in hexagonal close-packed crystals which normally slip only on the basal plane. Twinning can be said to result when a lamella within a crystal takes up a new orientation related to the rest of the crystal in a definite symmetrical fashion (see Figure A-23). The twin, thus, is a segment that has undergone a reorientation with respect to the parent crystal without change in crystal structure. The lattice within the twinned portion is a mirror image of the rest of the crystal. The plane separating the twin from the matrix is called the twinning plane. An axis perpendicular to the twinning plane is called the twinning axis. The twinning direction is the glide direction in the twinning plane of the atoms involved in the twinning process. The symmetrical relationship of a twin lamella and the matrix can be described as a rotation of

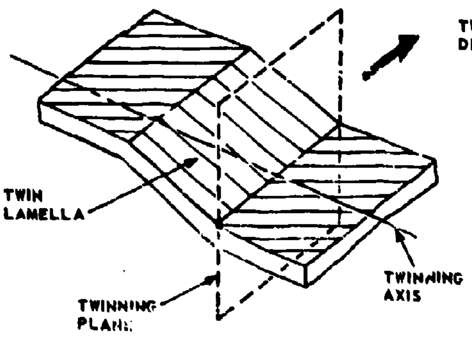


FIGURE A-23. ELEMENTS OF A TWIN LAMELLA WITHIN A GRAIN

the lamella through 180° about the twinning axis. It is not to be inferred that the atomic motion produces a twin. The atoms actually undergo some specific shearing translation over distances less than the interatomic spacing. The twin is described crystallographically by the twinning plane and twinning direction.

The crystallography of twinning is illustrated in Figure A-24. It should be noted that the distance an atom in the twin lamella moves is proportional to the distance from the twinning plane. The twinning process produces a disregistry on the free surface. On repolishing, the disregistry is removed but the twin can be discerned on etching because a permanent, specific reorientation of the lattice has occurred. The deformation that accompanies twinning leads to intense local stress in polycrystalline materials. The formation of a twin will be accompanied by slip-line formation in adjacent regions and may trigger the nucleation of twins in an adjoining grain. The regions at twin boundaries are favored positions for nucleation of recrystallization.

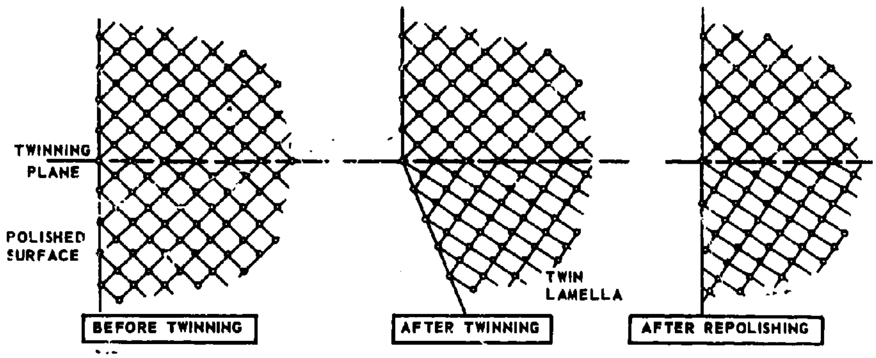


FIGURE A-24. CRYSTALLOGRAPHIC ASPECTS OF TWINNING

There are two classifications of twins according to the mode of formation, namely (a) deformation twins and (b) annealing twins. Deformation twins are those which develop as the result of mechanical loading. They form very rapidly and are usually thin lamellae within a grain. Their formation is favored by increasing the strain rate and by lowering the temperature, both of which mitigate against the slip process. Mechanical twinning occurs in body-centered cubic and close-packed hexagonal metals. The crystallography of these twins is summarized below:

Crystal system	Twinning plane	Twinning direction
Body-centered cubic	(112)	[111]
Close-packed hexagonal	$(10\bar{1}2)$	$[10\bar{1}\bar{1}]$

Face-centered cubic metals normally do not form deformation twins. They do, however, form twins on annealing after plastic straining, which accordingly are called annealing twins. They are believed to form concomitantly with grain growth, because the twin orientation can give a lower grain-boundary energy. Annealing twins are generally much broader than deformation twins. The twinning plane is (111) and the twinning direction is $[11\bar{2}]$ for face-centered cubic metals.

Deformation Bands. Many face-centered cubic metals and some body-centered cubic metals form bands on plastic deformation that resemble twins. The deformation bands differ from twin bands in that the orientation difference in a deformation band varies with the amount of straining, w' reas the twin immediately takes up a fixed-orientation . elationship with respect to the matrix. A grain begins to slio on a particular slip system. Eventually a band within a crystal prefers to slip on a different system. This event causes the band to rotate with respect to the grain and an orientation difference is produced which increases with increasing strain (see Figure A-25). The formation of deformation bands creates new interfaces which can offer resistance to slip and thereby harden the metal to a minor extent.

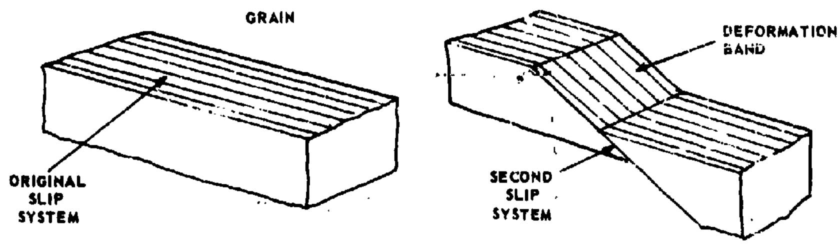


FIGURE A-25. FORMATION OF A DEFORMATION BAND

Deformation Mechanisms in the High-Temperature Range

The high-temperature range was defined previously as the range Tm/2 to Tm, where Tm is the melting point in degrees Kelvin. The temperature Tm/2 has been named the "equi-cohesive" temperature, below which the grain boundaries are more resistant to deformation by grain-boundary sliding than to slip inside the grain; the converse is true above this temperature.

Perhaps the most significant factor at Tm/2 and higher temperatures is that the rate of self-diffusion in metals becomes significant. This atomic movement takes place by exchange of atoms with vacancies. The exchange makes possible the phenomenon of "climb" of dislocations out of their slip planes, as illustrated in Figure A-26. Each exchange process lifts the dis. cation one interatomic distance out of the slip plane. This climb process allows dislocations to climb over obstacles and to recover from strain hardening. The rate of recovery increases with rising temperature because the rate of self-diffusion increases according to the well-known Arrhenius-type equation,

$$D = D_o^{(-Q/RT)} \tag{A-16}$$

where

D is the coefficient of self diffusion

D_o is a constant

Q is the activation energy

R is the universal gas constant

T is the temperature in degrees Kelvin.

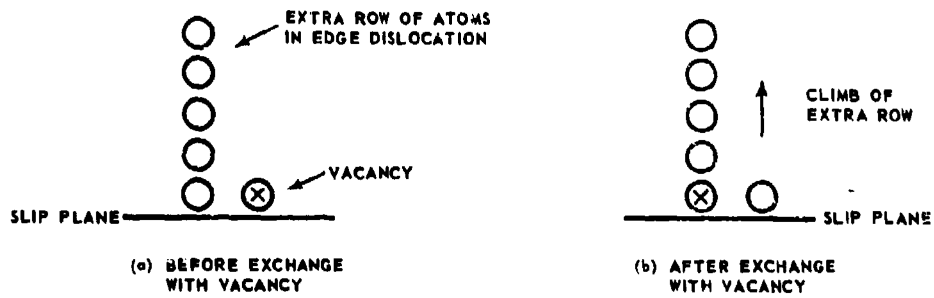


FIGURE A-26. CONCEPT OF CLIMB OF A DISLOCATION OUT OF THE SLIP PLANE

The processes occurring in the high-temperature range can be summarized as follows:

(a) Breakdown of grains into subgrains through dislocation climb
(b) Grain-boundary shearing and grain-boundary migration
(c) Fine slip, difficult to resolve by ordinary optical-microscopy techniques.

MACROSCOPIC DEFORMATION

The specific microscopic elements in the deformation of metals have been reviewed. The macroscopic manifestations of plastic flow in metals will be examined next.

Deformation of Single Crystals

It has been well established experimentally that deformation by slip begins when the resolved shear stress on the slip plane in the slip direction reaches a critical value, τ_C. The critical resolved shear stress is a function of alloy content, temperature, prior strain, grain size, and strain rate. Each slip system has a characteristic critical resolved shear stress. The resolved shear stress, depicted in Figure A-27, is given by

$$\tau_R = F/A \cos\lambda \cos\phi \quad . \qquad \text{(A-17)}$$

The condition for onset of slip is determined by the condition

$$\tau_R = \tau_C \quad , \qquad \text{(A-18)}$$

where τ_C is the critical resolved shear stress. For brittle materials (e.g., NaCl, Bi, Te, Zn), Sohnke's law is obeyed reasonably well in predicting the onset of cleavage fracture. It states that cleavage fracture will occur when the resolved normal stress (σ_R) reaches a critical value. The resolved normal stress is given by

$$\sigma_R = F/A \cos^2\phi \quad . \qquad \text{(A-19)}$$

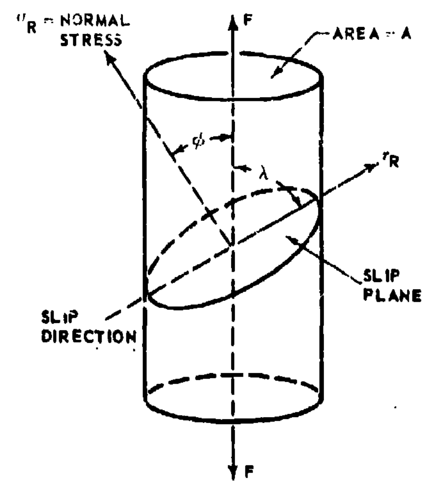


FIGURE A-27. THE RESOLVED SHEAR STRESS IN A SINGLE CRYSTAL IN SIMPLE TENSION

The critical condition for cleavage fracture is given as

$$\sigma_R = \tau_C \quad . \qquad \text{(A-20)}$$

Sohnke's law is not applicable to materials which exhibit significant capacity for plastic deformation.

It was previously stated that real solids contain viscous elements and that pure elastic response is expected only at absolute zero temperature and at zero load. The critical resolved shear stress is correctly interpreted as the shear stress at which the plastic strain rate becomes appreciable and easily detectable. In terms of the dislocation theory, the critical resolved shear stress, τ_C, is usually interpreted as the stress at which dislocation sources are activated.

<u>Rotation in Slip</u>. An important crystallographic feature of slip in metals is the crystal rotation accompanying this deformation mode. The slip plane rotates into the direction of the applied force in the case of tensile loading. In compressive loading, the rotation tends to orient the slip plane normal to the direction of the applied load. These rotations are thought to be the basis of preferred orientations or "textures" in the plastic deformation of polycrystalline metals. These deformation textures give rise to a "crystallographic" anisotropy of properties in metals in which the single crystal shows anisotropy. (The value of the property depends on the direction of testing.) A random texture is one in which every orientation in space has equal probability of occurrence as to all other orientations. When such a polycrystalline aggregate is tested for a particular property, it appears to be isotropic because a statistical averaging out of the property occurs.

The rotation of a crystal in the process of slip is demonstrated graphically in Figure A-28. For mathematical convenience, the glide plane is assumed to remain stationary, and the specimen axis to rotate. The specimen has length ℓ_0, before, and length ℓ_1, after deformation.

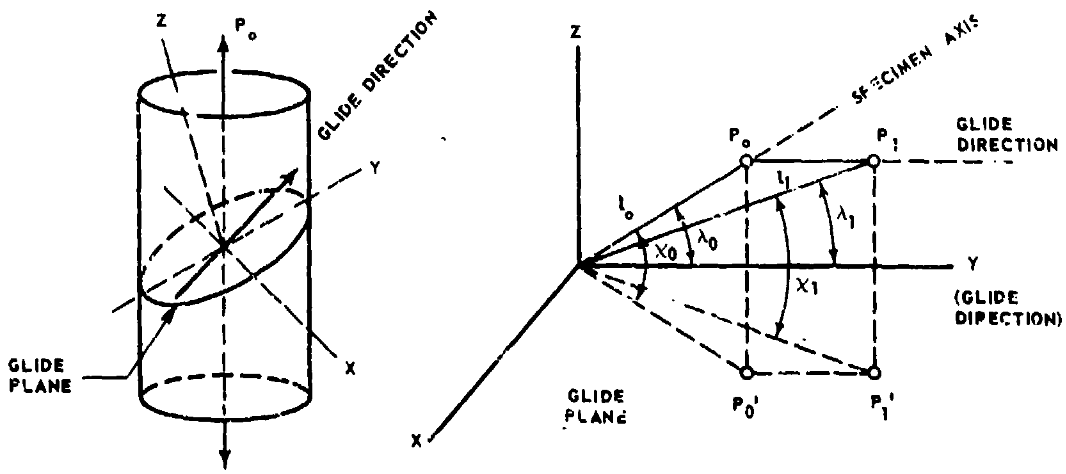


FIGURE A-28. ROTATION OF A CRYSTAL AS A RESULT OF THE SLIP PROCESS

In pure shear, each point $P_o(x_o, y_o, z_o)$ is transformed to $P_1(x_1, y_1, z_1)$ by gliding parallel to the Y-axis. In homogeneous shear, the displacement is proportional to the vertical distance z_o of P_o from the glide plane, leaving X- and Z-coördinates unchanged.

$$\left.\begin{aligned} x_1 &= x_o \\ y_1 &= y_o + z_o S \\ z_1 &= z_o \end{aligned}\right\} \quad \text{- - - -} \qquad \text{(A-21)}$$

Since

$$\ell_o^2 = x_o^2 + y_o^2 + z_o^2 \qquad \ell_1^2 = x_1^2 + y_1^2 + z_1^2$$

$$y_o = \ell_o \cos \lambda_o \qquad z_o = \ell_o \sin \chi_o \quad ,$$

it follows that

$$\frac{\ell_1^2}{\ell_o^2} = \frac{x_1^2 + y_1^2 + z_1^2}{x_o^2 + y_o^2 + z_o^2} = \frac{x_o^2 + y_o^2 + 2y_o z_o S + z_o^2 S^2 + z_o^2}{x_o^2 + y_o^2 + z_o^2}$$

$$= \ell + \frac{2 S y_o z_o + S^2 z_o^2}{\ell_o^2}$$

$$= \ell + 2S \cos \lambda_o \sin \chi_o + S^2 \sin \chi_o \quad . \qquad \text{(A-22)}$$

This equation enables one to calculate the amount of shear, S, from the extension in tension if the orientation of the crystal is known:

$$\epsilon = \text{extension} = \ell_1/\ell_o \quad .$$

Then

$$\ell_1{}^2/\ell_o{}^2 = 1 + 2S \sin\chi_o \cos\lambda_o + S^2 \sin^2\chi_o \quad ,$$

$$S^2 \sin^2\chi_o + 2S \sin\chi_o \cos\lambda_o + (1 - \epsilon^2) = 0 \tag{A-23}$$

and

$$S = \frac{-2 \sin\chi_o \cos\lambda_o \pm \sqrt{4 \sin^2\chi_o \cos^2\lambda_o - 4(1 - \epsilon^2) \sin^2\chi_o}}{2 \sin^2\chi_o} \tag{A-24}$$

and

$$S = \frac{1}{\sin\chi_o} \left(\sqrt{\epsilon^2 - \sin^2\lambda_o} - \cos\lambda_o\right) = \frac{\cos\lambda_1}{\sin\lambda_1} - \frac{\cos\lambda_o}{\sin\chi_o} \quad . \tag{A-25}$$

Equation (A-25) follows from

$$\epsilon = \ell_1/\ell_o \frac{\sin\chi_o}{\sin\chi_1} = \frac{\sin\lambda_o}{\sin\lambda_1} \quad ,$$

and so

$$\sin^2\lambda_o = \frac{\sin^2\chi_o}{\sin^2\chi_1} \cdot \sin^2\lambda_1$$

and

$$\frac{\sqrt{\epsilon^2 - \sin\lambda_o}}{\sin\chi_o} = \frac{\sqrt{\dfrac{\sin^2\chi_o}{\sin^2\chi_1} - \dfrac{\sin^2\chi_o \sin^2\lambda_1}{\sin^2\chi_1}}}{\sin\chi_o} = \frac{1}{\sin\chi_1} \sqrt{1 - \sin^2\lambda_1} = \frac{\cos\lambda_1}{\sin\chi_1} \quad .$$

It is evident from Equation (A-25) that the same amount of shear can give different extensions, depending on the orientation of the crystal.

The nature of the rotation of the crystal in slip is evident from the following argument:

$$z_1 = \ell_1 \sin\chi_o$$

since

$$z_1 = z_o = \ell_o \sin\chi_o \quad .$$

Then

$$\epsilon = \ell_1/\ell_o = \frac{\sin\chi_o}{\sin\chi_1} \quad ;$$

also

$$\ell_o \sin\lambda_o = \ell_1 \sin\lambda_1 \quad ,$$

so that

$$\epsilon = \ell_1/\ell_o = \frac{\sin\chi_o}{\sin\chi_1} = \frac{\sin\lambda_o}{\sin\lambda_1} \quad .$$

The angles χ_o and λ_o are fixed by the original orientation. As ϵ increases, it follows that χ_1 and λ_1 must decrease. Thus, the orientation of the crystal changes during extension so that the glide plane and glide direction approach the specimen axis.

Twinning. The deformation strains resulting from twinning in a crystal are represented in Figure A-29.[8] The configuration before straining is a sphere of unit radius. The upper half of the sphere is then distorted by a process of uniform shear into the ellipsoid that is the twinned position with respect to the lower half. The ellipsoid intersects the sphere in two circles in which the dimensions remain unchanged, one of which is the composition plane. The orientation of the second plane is related to amount of shear, S, where

$$S = \frac{B\,B^1}{\text{unit radius}}$$

or

$$S = 2 \cot 2\phi \quad \text{- - - - - -} \quad . \qquad \text{(A-26)}$$

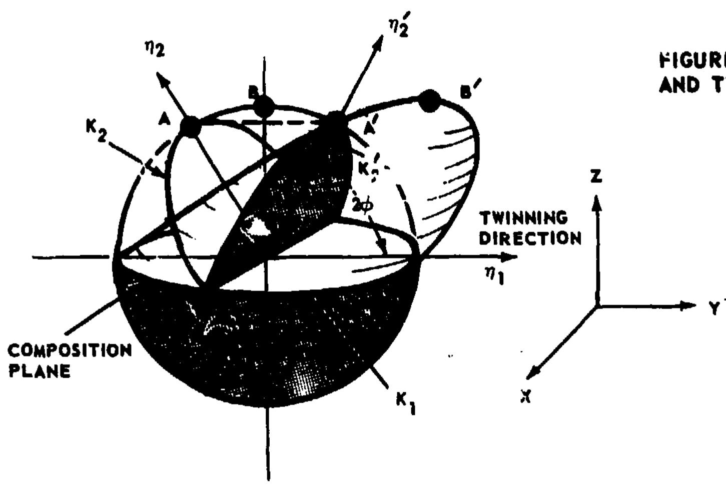


FIGURE A-29. RELATION OF SPHERE AND TWINNED ELLIPSOID

The composition plane is K_1 and the second undistorted plane is K_2'. The line through the center of the sphere, perpendicular to the line of intersection of K_1 and K_2' is the shear direction, η_1, and the plane perpendicular to K_1 and containing η_1 is the plane of shear. It intersects K_2 in a direction η_2. K_1 and η_2, or K_2 and η_1, are sufficient to determine the twinning.

The theoretical shears for body-centered and face-centered metals are 0.707. The theoretical shears for hexagonal metals depend on the c/a ratio. Some typical values are listed below:

Metal	c/a	Theoretical Shear (S)
Be	1.568	0.199
Ti	1.587	0.189
Mg	1.624	0.129
Zn	1.856	0.139
Cd	1.886	0.171

The extension in the crystal resulting from twinning is given by

$$\ell/\ell_o = \sqrt{1 + 2S \sin\chi \cos\lambda + S^2 \sin^2\chi} \quad , \tag{A-27}$$

where χ and λ are the angles between the specimen axis and the plane K_1 and the glide direction, η_1, respectively, in the undeformed crystal. The total extension resulting from twinning is small. For example, if a zinc crystal were entirely converted to a twin, the maximum elongation that would result would be 7.39 per cent. Twinning, however, can make an important contribution to the capacity for deformation in metals that have but one slip system, such as hexagonal metals. Twinning results in putting a part of the crystal in a new orientation which can be in a more favorable position for slip.

Deformation of Polycrystals

Most technologically important metals and alloys are fabricated and used in service as "polycrystalline aggregates", meaning that they are composed of a large number of "grains" or single crystals, each oriented differently in space relative to its neighboring grains. An extraordinary characteristic of such an aggregate of crystals is that it can be plastically deformed to a large degree and still maintain coherency among the grains.

Each grain has several faces bounded by a variable number of sides. The junction of three grains forms an edge. Plateau's rule states that four grain edges must be in a tetrahedral configuration, making angles of 109°28'. If the grain-boundary interfacial tension is equal throughout, dihedral angles of 120° should be established. The dihedral angle is the angle made by two grain faces on a reference plane normal to the edge.

An isolated crystal assumes a shape that minimizes its surface energy.[9] A crystal in a polycrystalline aggregate must fit perfectly into the hole formed by its neighbors, while minimizing its interfacial energy. Thus, its shape must satisfy the space-filling requirement along with minimization of interfacial tension. A crystal shape which meets these criteria is Kelvin's tetrakaidecahedron of minimum area, which has eight hexagonal and six square faces. Real metals, however, are found not to have this shape, presumably because the interfacial tensions are not equal with varying orientations of adjoining grains. Measurements on alloys show the number of sides per grain face varies from 3 to 8, the most frequently occurring face having 5 sides (pentagonal shape). Smith[10] gives the following relationship for polycrystalline aggregates:

$$C - E + F - B = 1 \quad , \tag{A-28}$$

where

C is number of corners

E is the number of edges

F is the number of faces

B is the number of grains.

From this expression, Smith predicts that the average number of sides per face should be 5-1/7.

McLean[11] states that, in the cold-working range, the main influence of grain boundaries is to increase the rate of strain hardening. The two important factors are that (a) the grain boundaries hinder slip and (b) the deformation within each grain is more complex than in an isolated single crystal. Polycrystalline metals are harder than isolated crystals because each crystal in the aggregate must deform in a complicated way in order to maintain coherency with its neighbors. The exact way in which a given grain accommodates to the deformation, change in shape, and rotation of its neighbors is not fully understood. Physical observations indicate that deformation is heterogeneously distributed within a grain. It seems reasonable that strain must be continuous across a grain boundary. Preferred orientations are generated by plastic deformation, indicating that grains rotate with respect to their neighbors.

The strength of polycrystals has been studied in several ways. Sachs[12] treats the problem as equivalent to loading a single crystal in a direction given by the average orientation of the grains in the polycrystal. The yield stress, τ, is given by

$$\tau = M_o T \quad , \tag{A-29}$$

where T is the critical resolved shear stress for the single crystal and $M_o = 2.238$, the orientation factor for a random orientation. Taylor[6] assumes that each grain undergoes essentially the same strain as the aggregate. An arbitrary change of shape of this type while a grain maintains coherency with its neighbors requires five operative slip planes. Taylor gives the expression for the yield stress, τ, as the following function of the critical resolved shear stress for the single crystal:

$$\tau = MT \quad , \tag{A-30}$$

where M is an accommodation factor equal to 3.06 and T is the critical resolved shear stress for a single crystal. Kochendörfer[13] considers the strength as made up of two components, the strength of each grain treated as an isolated grain, and the hardening from stresses near grain boundaries arising from the necessity of keeping contact. This second contribution is thought to be of minor importance in metals with many slip systems, but of major importance in metals with few slip systems.

Armstrong[14] states that two "size effects" should be considered in polycrystals; namely the "specimen size" effect and a "grain size" effect. These two effects, which must be differentiated, make additive contributions to strengthening.

As the number of grains in the section is increased, starting with a single crystal, the specimen-size effect is first to contribute to strengthening. This is due mainly to

the orientation dependence of crystal plastic flow. The following factors prevail: (a) the critical resolved shear stress must be achieved on the slip systems of all the grains contributing to bulk plastic flow and (b) the choice of slip systems in adjacent grains is affected by the relative orientation of the grains to each other, as well as to the applied stress, because strain continuity must be maintained at grain boundaries. This effect is independent of grain size.

The grain-size effect comes into play when the number of grains in the cross section is about 20 or more. Under these conditions, grain boundaries offer additional constraint to initiation and propagation of slip. The grain-size factor offers a barrier to flow which is proportional to the grain size but independent of grain orientation. The combined effect of the two constraints is given by

$$\tau = MT + Mk_s \ell^{-1/2} \quad - - - - \quad , \tag{A-31}$$

where

τ = shear stress for flow

MT = strength from specimen size

$Mk_s\ell^{-1/2}$ = strength from grain-boundary barriers (k_s is a constant, ℓ = grain size).

Armstrong's argument is summarized in Figure A-30.(14)

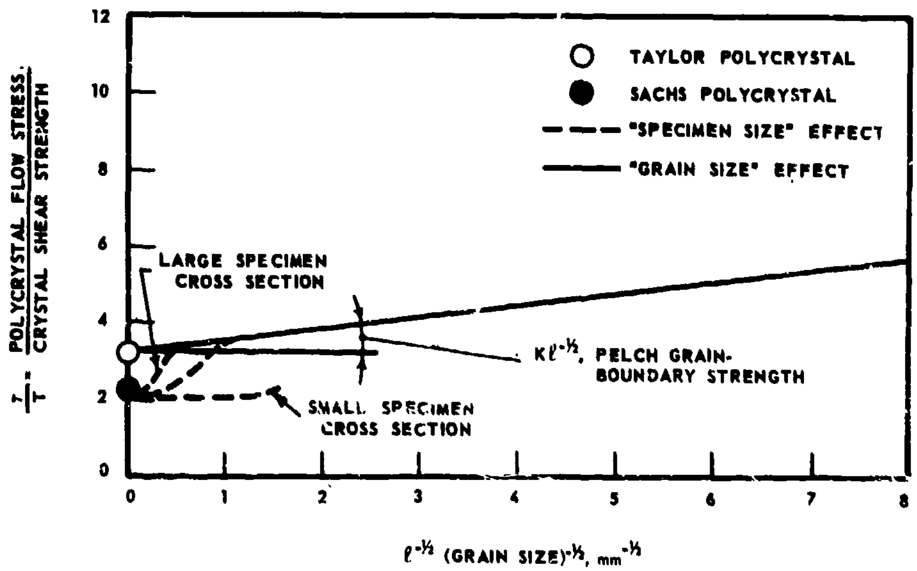


FIGURE A-30. FACE-CENTERED CUBIC POLYCRYSTAL FLOW STRESS AS FUNCTION OF GRAIN SIZE

TENSILE DEFORMATION

Let us now consider the response of polycrystalline aggregates to loads. Metals and alloys are generally fabricated and put into service as an aggregate of many crystals ranging in size from 20 to 250 microns (1 micron = 10^{-4} cm). When the individual grains, or crystals, are randomly oriented in space, a statistical "averaging out" of any property which is anisotropic in the single crystal is found when the aggregate is tested in any direction. This is why an aggregate simulates the classical homogeneous, isotropic body usually assumed in classical elasticity and plasticity. When a significant number of grains are aligned with a particular orientation in space, the anisotropy or directionality so developed must be taken into account.

The tensile test is one of the most useful quantitative tests of metals available in the laboratory. The test measures fundamental properties describing elastic response, flow characteristics, and onset of fracture. The specific properties obtained are the modulus of elasticity (stiffness), the flow stress (onset of plastic flow), the rate of strain hardening, and the fracture stress.

The two basic types of stress-strain curves encountered in tensile testing are the "engineer's diagram" illustrated in Figure A-31 for ductile metals and alloys which neck down prior to fracture.

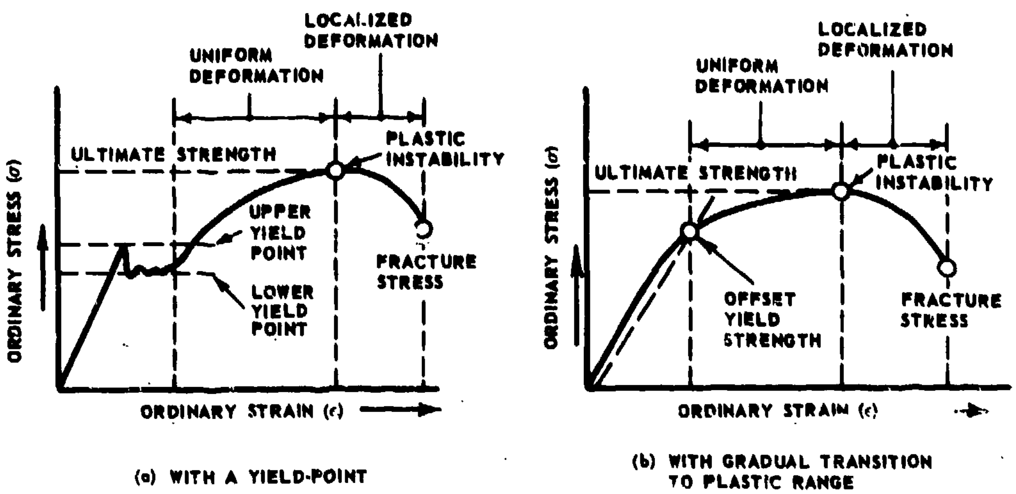
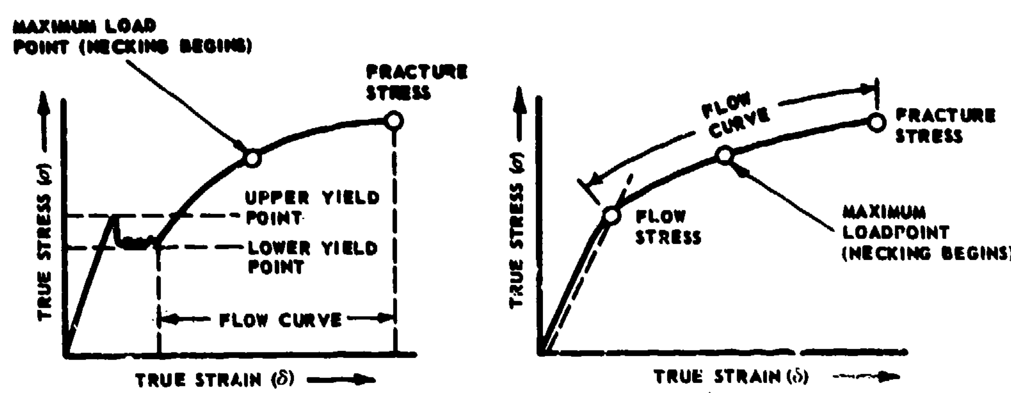


FIGURE A-31. TWO BASIC TYPES OF ENGINEER'S STRESS-STRAIN DIAGRAMS FOR METALS AND ALLOYS IN TENSION

For this type of plot, the stress is calculated as the load divided by the original cross-sectional area ($\sigma = F/A_0$) and the strain as the extension divided by the original gage length ($\epsilon = \Delta\ell/\ell_0$). Neither the cross-sectional area nor the gage length remain constant, and the plot is actually one of load versus extension divided by appropriate constants. Such a diagram is more accurately interpreted as a load-extension diagram. The maximum load point represents a condition of load instability for the particular

geometry of the test bar in which the balance between strain hardening and increase in stress owing to reduction in cross section is upset. At this point the bar starts deforming locally. This localized deformation introduces a notch in the specimen and the state of stress changes from uniaxial tension to unbalanced triaxial tension. The yield-point behavior is a manifestation of an overlooming phenomenon occurring principally in body-centered cubic metals (e.g., low-carbon steels), which is ascribed to the interstitial elements carbon and nitrogen. This is a special type of response, and the diagram in Figure A-31b should be considered the normal or expected tensile behavior for metals and alloys.

It is sometimes more useful, particularly in expressing plastic response, to deal with stress-strain relationships in terms of "true stress - true strain" diagrams. True stress is defined as the load divided by the instantaneous cross-sectional area ($\sigma = F/A$). The true strain is calculated as the summation of instantaneous strains $\left(\delta = \int_{l_0}^{l} \frac{dl}{l} = \ln \frac{l}{l_0}\right)$, which allows for a changing gage length. The two basic stress-strain diagrams plotted as true stress - true strain diagrams are shown in Figure A-32. The plastic section of the curves beyond the flow stress now rises continuously to fracture. This portion of the curve is called the "flow curve", because each point on the curve represents a flow stress for a specimen with that particular strain history. There is no discontinuity in the flow curve at the maximum load point (onset of plastic instability). Thus, on this basis, a metal or alloy should be regarded as having two basic strength characteristics, namely, the resistance to plastic flow (flow stress) and the resistance to fracture (fracture stress). It is doubtful that metals have unique fracture stresses, but nevertheless it is useful to record the true fracture stress. A metal that shows continuously rising stress to fracture is said to exhibit strain-hardening characteristics.



(a) WITH A YIELD POINT

(b) GRADUAL TRANSITION TO PLASTIC FLOW

FIGURE A-32. TWO BASIC STRESS-STRAIN CURVES PLOTTED AS TRUE STRESS VERSUS TRUE STRAIN

Two ductility figures are generally recorded in the tensile test, and are interpreted as follows: the elongation is a measure of the ability of the specimen to flow uniformly, whereas the reduction of area is a measure of its ability to undergo localized deformation.

It is found empirically, for many metals and alloys, that the flow curve (from the flow stress to the fracture stress) can be expressed by the relationship

$$\sigma = K\delta^{n} \quad , \qquad \text{(A-32)}$$

where

σ is the true stress

K is the strength coefficient

δ is the true strain

n is the strain-hardening exponent.

Thus, a plot of the logarithm of the stress versus the logarithm of the strain will yield a straight line, if a correction is made for the stress resulting from the triaxial state of stress that accompanies the phenomenon of necking. Another interesting feature of the flow curve is that the true strain at the maximum load point is equal to the strain-hardening exponent ($\delta_{MLP} = n$). The strain-hardening exponent thus gives the magnitude of the uniform deformation in tension before necking or instability sets in.

The rate of strain hardening at any value of the strain is given by differentiation of Equation (A-32):

$$\frac{d\sigma}{d\delta} = Kn\delta^{n-1} \quad . \qquad \text{(A-33)}$$

Variables Influencing the Flow Curve

The properties of the flow curve (flow stress, strain-hardening characteristics, and fracture stress) depend on the following variables: (a) metallurgical structure, (b) prior strain history, (c) state of stress, (d) strain rate, and (e) temperature. The effect of some of these variables will be discussed here briefly; variables (a) and (b) will be discussed later. In general, these variables influence the flow stress more than the fracture stress.

State of Stress. The state of stress is defined as the direction and magnitude of the components of stress at a point in a body under load. If one considers an infinitesimal cube at this point, all the forces acting on the cube can be resolved on any of its faces as one normal and two shear stresses, yielding 18 components of stress. These can be reduced to three normal and three shear stresses. The cube, however, can be rotated into a unique orientation in which (1) the three normal stresses are maximized and (2) the shear stresses on the faces of the cube go to zero. The cube axes are now in the principal triaxial directions and the normal stresses are called the principal normal stresses σ_1, σ_2, σ_3 ($\sigma_1 > \sigma_2 > \sigma_3$ in the algebraic sense). The state of stress is usually expressed in terms of the principal normal stresses. The stresses of particular interest are σ_1, the maximum normal tensile stress, and $\tau_{max} = \frac{\sigma_1 - \sigma_3}{2}$, the maximum shear stress. The normal stress can lead to cleavage fracture and the shear stress can lead to plastic flow. The state of stress is often referred to in terms of the ratio σ_{max}/τ_{max}.

Lankford, Low, and Gensamer[15] derived expressions for the flow in biaxial and triaxial tension in terms of the flow stress in simple tension. For biaxial stress states, they obtain

$$\sigma_1 = K\delta_1^{\,n} \frac{2^n}{(\alpha^2 - \alpha + 1)^{1-n/2}\,(2 - \alpha)^n} \quad , \tag{A-34}$$

where

σ_1 is the maximum principal normal stress

K is the strength coefficient

δ_1 is the maximum principal normal strain

n is the strain-hardening exponent

α is the stress ratio σ_2/σ_1 ($\sigma_3 = 0$) .

For triaxial stress states, the expression becomes

$$\sigma_1 = K\delta^n \frac{2^n}{(\alpha^2 + \beta^2 - \alpha\beta - \alpha - \beta + 1)^{1-n/2}\,(2 - \alpha - \beta)^n} \quad , \tag{A-35}$$

where β is the stress ratio σ_3/σ_1.

The effect of the state of stress on the flow curve is illustrated in Figure A-33 for several states of stress. Biaxial states raise the curve to a modest extent but not successively with increasing values of α. In triaxial tension, the flow curve is significantly

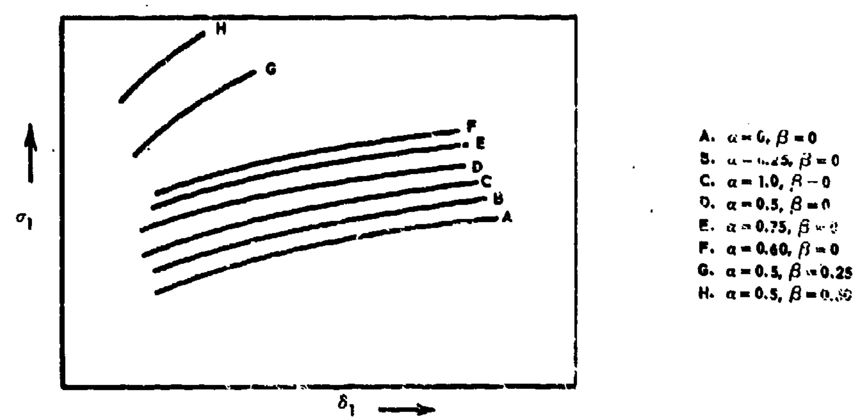


FIGURE A-33. EFFECT OF STATE OF STRESS ON THE FLOW CURVE IN TENSION

raised and the plastic flow prior to fracture is significantly decreased. This accounts for the typical effect of a notch, which sets up a triaxial stress state.

Strain Rate. It is generally known that increasing the strain rate or rate of loading raises the flow stress and the flow curve. This mode of decreasing the capacity for plastic flow is not so effective, however, as introducing a triaxial tension state of stress or lowering the temperature. The flow curve is raised some 15 to 30 per cent for most metals and alloys below their recrystallization temperature, and somewhat less than this at lower temperatures. Zener and Hollomon[16] determined experimentally that the logarithm of the stress to produce a given amount of plastic strain is linearly related to the logarithm of the strain rate:

$$\sigma_\epsilon = K\left(\frac{d\epsilon}{dt}\right)^n \quad , \tag{A-36}$$

where

σ_ϵ is the stress required to give strain

K and n are temperature-dependent constants

$\frac{d\epsilon}{dt}$ is the strain rate.

The coefficient n usually falls rapidly with temperature. For example, n for copper is 0.12 at 600 C and 0.01 at -253 C.[17]

It is important to note another effect of strain rate on the flow stress. The relaxation of stored elastic energy through the process of plastic flow is almost entirely manifested as sensible heat. If the strain rate is sufficiently high that the heat generated cannot be removed, the plastic flow takes place adiabatically, with a sharp temperature rise. Because the rise in temperature leads to a decrease in flow stress, increasing the strain rate may not produce the expected increase in flow strength.

Temperature. Lowering the temperature is an effective way of raising the flow curve. For metals and alloys which recrystallize above ambient temperature, the flow stress and flow curve can be raised by a factor of two by lowering the temperature from ambient to -253 C, and by a factor of four for metals and alloys which recrystallize at ambient temperature (e.g., lead).[18]

Hollomon[19], from an analysis of a considerable amount of experimental data, proposed the relationship

$$\sigma_\epsilon = Ke^{Q/RT} \quad , \tag{A-37}$$

where

σ_ϵ is the stress at some strain

K is a constant

T is the absolute temperature

R is the gas constant

Q is a parameter depending on the material and perhaps the temperature range,

to predict the effect of temperature on the flow curve.

Expressions for the dependency of τ_c, the critical resolved shear stress for the onset of plastic flow, on temperature for metal single crystals are given by Orowan and Meacham. Orowan[20], using a dislocation model for slip and taking into account the contribution of thermal vibrations of the lattice, derives the relationship

$$\tau_c = A - B\sqrt{T} \quad , \tag{A-38}$$

where A and B are constants and T is absolute temperature. This expression was found to agree with the behavior of zinc and cadmium single crystals with proper choices of A and B.

Meacham[21] derived an equivalent parabolic expression

$$\tau_c = \phi(\mu, T) = \tau_0\left(1 - \frac{k}{A_1} \ln \frac{\alpha_1\lambda}{\mu} \sqrt{T}\right) \tag{A-39}$$

where

τ_0 is the critical shear stress at absolute zero

k is Boltzmann's constant

A_1 is the thermal energy required for an atom to cross the potential-energy barrier to form a dislocation

α_1 is a proportionality constant

λ is the interatomic spacing

μ is the relative strain rate.

Any relationship between flow stress and temperature requires that, in the temperature range considered, no reactions occur that can lead to a change in metallurgical structure, such as precipitation, strain aging, and recrystallization.

Equivalence of Strain Rate and Temperature

It has been recognised for some time that increasing the rate of straining is qualitatively equivalent to lowering the temperature, with respect to the response of metals to applied loads. Both changes are found to raise the flow curve and in general reduce the capacity for plastic flow before fracture. Holloman and Zener[22] proposed a quantitative expression of this equivalence in the form

$$p = \dot{\epsilon} e^{Q/RT} \tag{A-40}$$

where

p is a parameter

$\dot{\epsilon}$ is the strain rate

Q is the activation energy

R is the gas constant

T is the absolute temperature.

The flow curve will be identical for any combination of strain rate and temperature that gives the same value of the parameter p. This relationship predicts the behavior of metals reasonably well, provided the rang of temperature or strain rate is not too large and Q can be taken as constant.

Equation (A-40) permits the calculation of a strain rate that is equivalent to a specific temperature change, as follows

$$\dot{\epsilon} \cdot e^{Q/RT_r} = \dot{\epsilon}_r \cdot e^{Q/RT}$$

or (A-41)

$$\dot{\epsilon} = \dot{\epsilon}_r \cdot e^{Q/R\left(\frac{1}{T} - \frac{1}{T_r}\right)} \quad ,$$

where the subscript "r" denotes room-temperature properties. These relationships do not have universal application and should be used with discretion.

SIMPLIFIED MODELS OF FLOW AND FRACTURE

A simple qualitative picture of conditions of flow and fracture of metals can be based on two strength parameters of the material; namely, τ_c, the critical resolved shear stress for plastic flow, and σ_c, the critical normal stress for fracture, and on two stress parameters which depend on the manner of loading (state of stress); namely $\sigma_1 = \sigma_{max}$, the maximum normal stress, and $\tau_{max} = \frac{\sigma_1 - \sigma_3}{2}$, the maximum shear stress. The metal will show ductile behavior if τ_{max} reaches the value of τ_c before σ_{max} reaches the critical value σ_c, and brittle behavior if the converse occurs. Gensamer[23] has depicted this model as shown in Figure A-31. This model is over-simplified, but is a useful qualitative argument for predicting the expected degree of ductility for a particular set of circumstances.

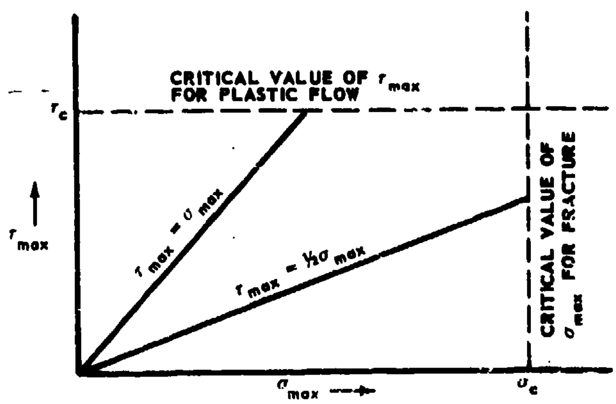


FIGURE A-31 ROLE OF RATIO τ_{max}/σ_{max} IN DETERMINING DUCTILITY

An important consideration in the fabrication of body-centered cubic metals and alloys is the fact that they undergo embrittling transitions as the test temperature is lowered. Such a transition in impact strength is illustrated schematically in Figure A-35, along with the typical behavior of face centered cubic and hexagonal close-packed metals. Neither face-centered cubic nor hexagonal close-packed metals undergo embrittling transition, but the toughness of the former structure is much greater. The impact-energy value can be rationalized in terms of the volume undergoing plastic deformation (V_d) and the energy absorbed per unit volume, which is the area under the tensile stress-strain curve. The embrittling transition represents a loss of ability to propagate plastic strain in V_d. This transition can occur from well below ambient temperature to well above it. The transition temperature can be particularly high for the refractory metals (Mo, W, and Cr).

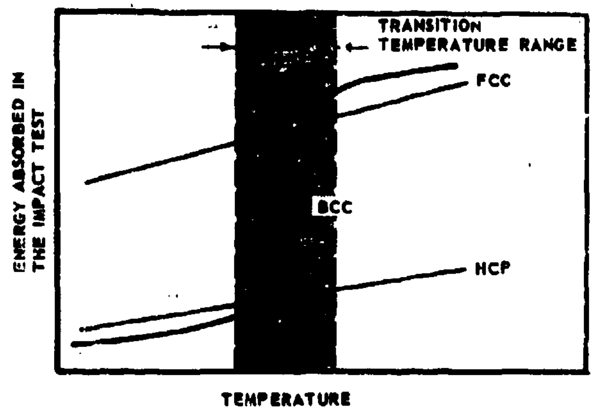


FIGURE A-35. SCHEMATIC REPRESENTATION OF THE IMPACT ENERGY-TEMPERATURE RELATIONSHIP FOR THREE PRINCIPAL METAL LATTICE STRUCTURES

The body-centered cubic metals undergo embrittlement at successively lower temperatures in impact, tension, and torsion, for which the ratios of σ_{max}/τ_{max} are > 2.0, 2.0, and 1.0, respectively. The order of embrittlement is rationalized on the basis of the marked temperature dependence of the flow stress and the relationships shown in Figure A-34. It is important to note that the transition temperature for brittle fracture depends on testing conditions because it is influenced by the loading rate and the stress system.

The process of plastic flow in metals is terminated by one of several modes of fracture, causing the separation of a body into two or more parts and the creation of new surfaces. This is a natural and basic limitation of a forging process, as is the lack of sufficient press capacity in overcoming the flow stress of the material. The fracture stress is much less sensitive to certain factors than is the flow stress. Increases in strain rate or prior strain and decreases in temperature increase the resistance to flow, but have relatively little effect on the fracture stress.

The idealized concept of fracture is the pulling apart of a layer of atoms from the neighboring layer, requiring the overcoming of the bonding forces between the atoms. The theoretical strength is calculated by equating the energy required to separate atoms

an infinite distance to the energy required to form two new surfaces. The expression derived in this manner yields the following relationship:

$$\sigma_{th} = \sqrt{\frac{ES}{a}} \qquad \text{(A-42)}$$

where

σ_{th} is the theoretical strength

E is the modulus of elasticity

S is the surface energy per unit area

a is the atomic spacing.

Fracture stresses calculated by this relationship yield values that are 100 to 1000 times higher than the observed values. Obviously, fracture in metals and alloys does not occur by this idealized mechanism. It should be considered, rather, that fracture is a result of the process of plastic flow, and that plastic action is a necessary stage for the development of fracture.

It is convenient to discuss fracture mechanisms in terms of single crystals and polycrystalline aggregates.(7) It should be understood, however, that seldom, if ever, will fracture occur as the result of a single mechanism.

Fracture of Single Crystals

(1) Shear Fracture

Single crystals of metals such as magnesium will deform by slip, the sliding of one part of the crystal over another. Shear fracture is said to have occurred when this sliding is continued to a point at which the two parts of the crystal are completely separated. Thus, shear fracture is the termination of the slip process.

(2) Glide, or Glide Plus Twinning (No Definite Fracture Stage)

Examples of this behavior are found in single crystals of zinc or cadmium. Slip starts on the basal plane, twinning occurs on the (1012) plane, which is followed by secondary basal slip in the twin bands. This deformation mode produces necking of the specimen, and the specimen draws to a wedge and separates along a line. There is no fracture stage.

(3) Cleavage

At low temperatures, single crystals may fracture by cleavage, or separation perpendicular to some crystallographic plane, producing a bright, smooth surface. This type of fracture occurs on distinctive crystallographic planes. No face-centered cubic metal is known to fail by this mechanism.

Fracture of Polycrystalline Metals

(1) Metals of Moderate Ductility

Metals and alloys such as copper, aluminum, and steel (depending on the strength level) undergo considerable plastic flow and necking down. Fracture initiates within the specimen and propagates with increasing plastic strain, gradually opening up the fracture toward the surface of the specimen. The mechanism finally changes to shear, yielding a "cup-cone" fracture. In some cases, a 45-degree fracture surface is obtained instead of the cup cone.

(2) Drawing to a Point (100 Per Cent Reduction of Area)

Metals such as fine-grained zinc, lead, or gold draw down to a point and then separate; there is no definite fracture phase. This type of separation is promoted by applying hydrostatic pressure.

(3) Low Temperature

(a) Those metals which cleave as single crystals will cleave as polycrystalline aggregates. In cleavage, there is no apparent increase in strain to propagate the fracture.

(b) Brittle intergranular fracture at ambient temperature can occur because of structural or compositional irregularities. Examples are overheating (burning or oxidation) and temper brittleness in steel.

(4) Elevated Temperatures

The fracture changes from transcrystalline to intercrystalline above the equicohesive temperature range.

(5) Alternating Stress

Metals subjected to repeated loads fracture by two general mechanisms. At stresses corresponding to the fracture curve, rupture occurs when the ductility is exhausted. Fracture at the endurance limit occurs as a result of formation of disordered layers to a certain depth and the joining up of these layers by cracking, followed by general disintegration.

(6) Stress-Corrosion Cracking

The phenomenon known as stress-corrosion cracking is fundamentally a corrosion process accelerated by stresses, either internally or externally applied. The stresses store up excess strain energy, and aid in breaking protective films. The cracking may be promoted by the wedge effect of solid corrosion products.

Fracture Nomenclature and Mechanisms

The nomenclature for fracture phenomena is varied and confusing because of the many types of fracture encountered in metals. The terms generally arise from description of the crystalline mode of fracture, appearance of the fracture surface, and the

amount of plastic strain preceding fracture. The common system(23) of terms may be summarized as follows:

Behavior Described	Descriptive Terms
Crystalline mode	Shear cleavage
Appearance of fracture	Fibrous granular
Amount of strain to fracture	Ductile brittle

For example, a truly ductile specimen breaks in the shear mode, and the fracture surfaces exhibit a fibrous appearance or texture.

It is useful to classify fractures in real polycrystalline metals and alloys under two general categories of ductile and brittle fractures.

Ductile Fracture. Plastic strain is inherent in ductile fracture, including the initiation, the propagation, and the final stage of separation. Lattice vacancies are produced by plastic flow, leading to pore formation by vacancy condensation. These pores are extended and opened through the action of plastic flow, which is an inherent part of the fracture process. There is no definite fracture stage as in the development of a running brittle fracture. The concept that a certain amount of plastic flow sets up a criterion of rupture has proved to be incorrect.

Brittle Fracture. The initial requirement in brittle fracture is a crack nucleus of critical size that will develop into a running crack. This critical-size crack nucleus may arise from pre-existing flaws, may form by the ductile-fracture mechanism through the action of plastic flow, or may result from the joining up of two existing subcritical-size flaws. The critical stage in brittle fracture is the transition from a slowly propagating crack to a fast-running crack. This sharp increase in propagation velocity greatly limits the dissipation of stored-up energy into plastic flow.

Role of Inclusions in Ductile Fracture. Most forging operations involve considerable amounts of deformation, so that ductile fracture is of particular interest. Brittle fracture may be induced in some cases by impact loading.

It will be recalled that ductile fracture involves the formation of pores and the extension and joining of pores by plastic flow, leading to ultimate separation. Pores may form by the following mechanisms:

(1) Formation of vacancies by concentrated plastic flow, coalescence of vacancies into pores, and extension of pores by plastic strain.

(2) Nonmetallic inclusions as nucleation sites. Two general cases are of interest:

(a) DECOHESION

(b) CAVITY IN INCLUSION

Experimental evidence shows that the presence of nonmetallic inclusions has an important effect on the tensile reduction of area, which is a direct measure of total strain before fracture. For example, high-purity aluminum has reduction-of-area values of 90 per cent, whereas the typical value is 30 per cent for impure commercial grades of aluminum. Oxide inclusions in copper considerably reduce its ductility. Thus, inclusions can considerably limit the plastic flow of metals. The minimization of inclusions is one of the expected benefits of vacuum melting.

RELATIONSHIP BETWEEN ELASTIC AND PLASTIC PROPERTIES OF PURE POLYCRYSTALLINE METALS

The relationships between stress and strain in the plastic state are complicated by several factors, including the following:

(1) The magnitude of plastic strains is appreciable and the usual practice of dropping infinitesimals of an order greater than one in derivations of relationships introduces serious errors,

(2) There is no simple relationship between stress and strain, and

(3) The principal strains do not necessarily remain parallel to the principal stresses.

Pugh[24] has analysed the possibilities of relationships among parameters of plastic deformation and the elastic constants for pure polycrystalline metals. The basic premise of the treatment is that most of the energy of a dislocation is stored outside the core, in that part of the lattice where the strains are Hookean. The elastic constants are related to (a) the resistance to plastic deformation, (b) the fracture strength, and (c) the extent of the plastic range, or the malleability.

The expression for the hardness, S, which is indicative of the flow stress, is derived from a dislocation model using a Frank-Read source. The flow stress, S, is that stress required to activate a Frank-Read source of length ℓ, or

$$S \sim G \cdot b/\ell \quad , \qquad \text{(A-43)}$$

where

G is the rigidity modulus

b is the Burgers vector of the dislocation.

The length, ℓ, is assumed to be constant for all metals, yielding an expression for "softness" (reciprocal of the hardness) as follows:

$$\frac{1}{B} = \frac{c}{G \cdot b} \quad , \qquad \text{(A-44)}$$

where c is a constant for all metals of the same structure.

The "softness" of a metal is assumed to consist of two components, the first due to movement of dislocations expressed by Equation (A-44), and the second due to thermally activated processes. The latter component is a function of temperature and rate of straining. The two components yield the relationship

$$\frac{1}{B} = \frac{c}{G \cdot b} + f(T/T_m, t) \quad , \qquad \text{(A-45)}$$

where

T/T_m is the reduced, or homologous temperature (T_m is the absolute melting point), and

t is the time of stressing.

For metals melting at temperatures above 1000 C and tested at ambient temperature, $f(T/T_m, t) \simeq 0$, and the constant "c" can be calculated. To test the contribution of the thermally activated processes to the softness, the factor G·b/B is plotted against the melting point for the three main lattice structures of metals in Figure A-36. The parameter is constant for metals with melting points above 900 C, but rises steeply for those melting below this temperature. The anomalous behavior of Ca and Pt is usually rationalised on the basis of purity and anisotropy. The body-centered metals can be divided into two subgroups, the low-melting group including the lithium and sodium group, and the high-melting group starting with iron. The value for W is in doubt because of uncertainties about purity. The hexagonal metals also can be divided into two groups, melting above and below 900 C. Pugh discusses the discrepancies from various viewpoints.

The temperature dependence of the elastic moduli of metals is similar from absolute zero to about $T_m/3$. Above this temperature, thermally activated processes enter in and the $f(T/T_m, t)$ parameter must be determined and taken into account. Thus, the prediction of hardness (or softness) of metals at ambient temperature by Equation (A-44) is restricted to metals melting above 900 C. The approximate values of the constant C are the following: face-centered cubic - 0.33; body-centered cubic - 0.27; and hexagonal - 0.16.

The intrinsic fracture stress is estimated from the area under the force-displacement curve as two atoms are separated to an infinite distance. The abscissas of this curve are proportional to the lattice parameter, a, and the ordinates to the bulk modulus, K. The intrinsic fracture stress, σ, is then proportional to K·a. Pugh argues that, to relate the intrinsic stress to true fracture stress, it is necessary to take into account the stress concentration at the base of the crack. It is argued that the crack becomes less sharp as plastic shear stress decreases, giving the expression for true fracture stress,

$$\tau \sim K \cdot a / f(G \cdot b / K) \quad . \qquad \text{(A-46)}$$

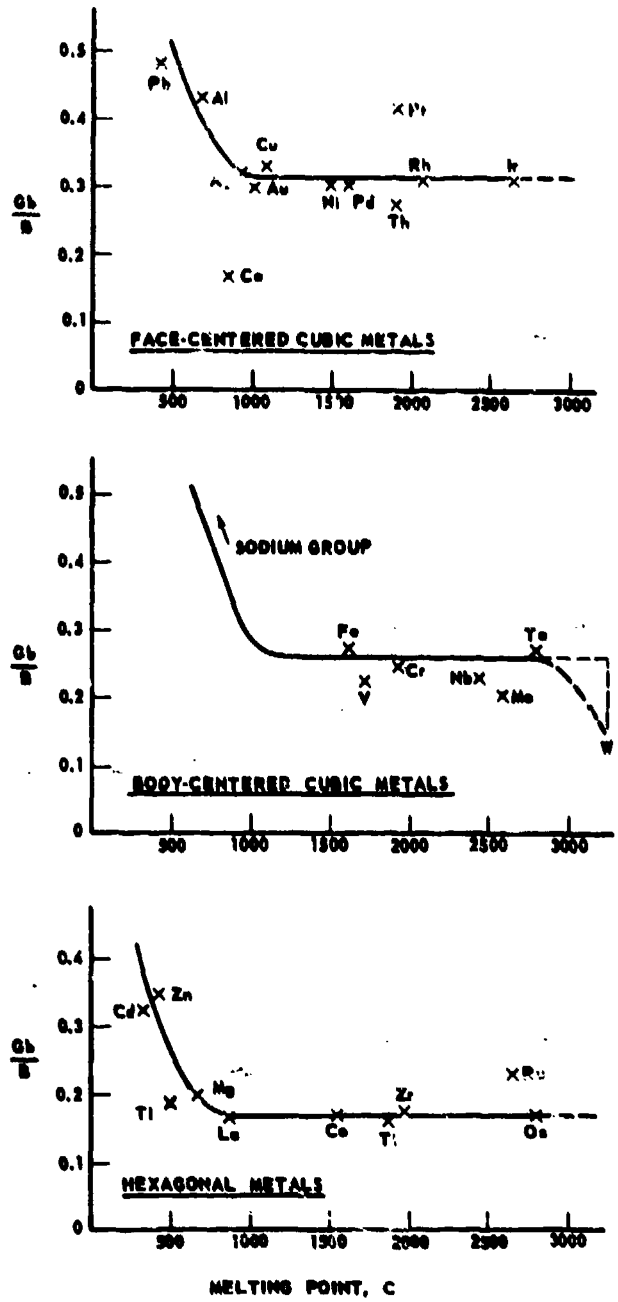


FIGURE A-36. TEST OF CONTRIBUTION OF THERMALLY ACTIVATED PROCESSES TO "SOFTNESS" OF METALS

It is further argued that the stress-concentration factor is small and constant among metals, giving as a first approximation

$$\tau \cdot \; K \cdot a \quad , \tag{A-47}$$

The malleability (plasticity) of a metal is predicted from the gap between the flow stress and the fracture stress. The treatment is based on the true stress – true strain curve, assuming that the stress for any given strain is given by G·b·g and the strain-hardening function, where g is constant within each group of metals. By a series of arguments, it is deduced that plastic strain, ϵ_p, is given as

$$\epsilon_p = \phi(K/G) \quad . \tag{A-48}$$

The higher the ratio of the bulk modulus to the shear modulus, the greater the malleability. Since the Poisson's ratio, ν, is related to K/G, the malleability should also be related directly to this elastic constant. The values of K/G and ν are listed for face-centered cubic and hexagonal metals in Table A-2. The body-centered cubic metals all have approximately the same values of K/G and ν, indicating these metals should have comparable malleabilities.

COLD WORKING AND HOT WORKING

It is important in forging processes to distinguish between cold working and hot working and the role of the process of recrystallisation in this distinction. Cold working is considered to be plastic deformation accomplished below the recrystallisation temperature range and hot working, that done above this temperature range. The term "warm working" is used to designate plastic deformation at elevated temperatures but below the recrystallisation temperature.

Cold Working

Cold working is characterised by strain hardening, which is one of the basic mechanisms of hardening metals. Both ultimate strength and yield strength increase as the amount of strain decreases, whereas the ductility generally decreases.

The details of strain-hardening mechanisms in metals have yet to be clarified. It is certain that the density of defects, particularly dislocations, is increased by cold working. Also, the internal surface per unit volume (grain boundaries) increases, and elastic strain energy is stored because of lattice bending, particularly at grain boundaries. These increments of energy stored in cold-worked metals constitute the "latent heat of cold work" which is given up when the metal is softened by annealing. Less than 10 per cent of the total work of deformation is stored up in the metal (1 calorie or less per gram); the rest of the work of deformation is given off as sensible heat.

The latent heat of cold work or extra stored energy in the metal is the basis of the following characteristics of the cold-worked state:

(1) The internal energy is increased by cold work
(2) The density decreases slightly
(3) The thermal coefficient of expansion increases slightly

TABLE A-2. CRITERION K/G AND ELONGATION FOR FACE-CENTERED CUBIC AND HEXAGONAL CLOSE-PACKED METALS

Metal	Rigidity Modulus, G, Kg/mm^2	Bulk Modulus K, Kg/mm^2	K/G	Poisson's Ratio	Elongation, per cent	Purity, per cent
			Face-Centered Cubic			
Calcium	750	1750	2.33	0.31	60	
Strontium	620	1220	1.97	0.28		
Aluminum	2720	7460	2.74	0.34	50	99.99
Thorium	3160	5500	1.74	0.26		
Nickel	7500	18900	2.52	0.32	30	99.95 (Nitco)
Rhodium	15300	27000	1.77	0.26	Small	
Iridium	21400	37300	1.74	0.26	Small	
Palladium	4450	19000	4.27	0.39	40	99.9+
Platinum	5300	27800	5.25	0.44	40	
Copper	4640	13900	3.00	0.35	60	99.99
Silver	2940	10100	3.44	0.37	60	99.98
Gold	2820	17300	6.14	0.42	50	99.99
Lead	570	4200	7.37	0.44	64	99.99
			Hexagonal Close-Packed			
Beryllium	13500	11700	0.867	0.08	1	
Magnesium	1770	3390	1.915	0.28	16	99.98
Lanthanum	1500	2840	1.90	0.28		
Titanium	3870	12500	3.23	0.36	37	
Zirconium	3540	9100	2.58	0.33	40	
Hafnium	3100	11100	3.58	0.37		
Cobalt	7630	18500	2.43	0.31		
Rhenium	21000	37000	1.76	0.26		
Ruthenium	17600	29000	1.65	0.25		
Osmium	22800	38000	1.67	0.25		
Zinc	3790	6000	1.59	0.27	25	99.99
Cadmium	2460	5000	2.03	0.29	50	99.99
Thallium	280	2900	10.3	0.45	Large	

(4) The elastic moduli are not greatly affected
(5) Cold work increases the reactivity of metals.

Thus, cold-worked metals are in a metastable condition because of the extra stored energy, and tend to relax to an equilibrium state of lowest free energy. The rate of relaxation to a lower energy state is dependent exponentially on the temperature.

Larson and Kula[25] examined the influence of crystal structure on the strain-hardening characteristics of high-purity polycrystalline metals. The effect of crystal structure on the strain-hardening exponent, n, in the expression $\sigma = K\delta^n$ for the three principal crystal structures are summarized in Table A-3. These are mean values of "n" based on data over a temperature range in which there is little change in the true strain at the maximum load point. The data in the table fit the relationship

$$n = k(SD) \quad , \tag{A-49}$$

where

k is a constant, about 0.06

SD = number of slip directions.

TABLE A-3. INFLUENCE OF CRYSTAL STRUCTURE ON STRAIN HARDENING

Crystal Structure	True Strain(a) at Maximum Load	Number of Slip Directions	Number of Slip Systems
Hexagonal close packed	0.18	3	3
Body-centered cubic	0.26	4	48
Face-centered cubic	0.38	6	12

(a) n (strain-hardening exponent).

It is found that the parameter n increases as the grain size increases. Larson and Kula derive the relationship

$$k = K(GS)^S \quad , \tag{A-50}$$

where

k = k in Equation (A-49)

K = a constant (the intercept at a grain size of 1 mm, equal to about 0.18)

GS = grain size in millimeters

S = slope of the log n - log grain size plot (about 0.18).

The effect of temperature on n is quite different for different crystal structure, as shown in Figure A-37. In all cases, the capacity for strain hardening is lost as high temperatures are attained. As absolute zero is approached, the face-centered cubic metals show a significant increase in n; the hexagonal close-packed metals show an even greater increase because of the onset of twinning at low temperatures. The decrease in n for the body-centered cubic metals as the temperature approaches absolute zero is unique. This behavior may be a basic factor in the embrittling transition exhibited by these metals.

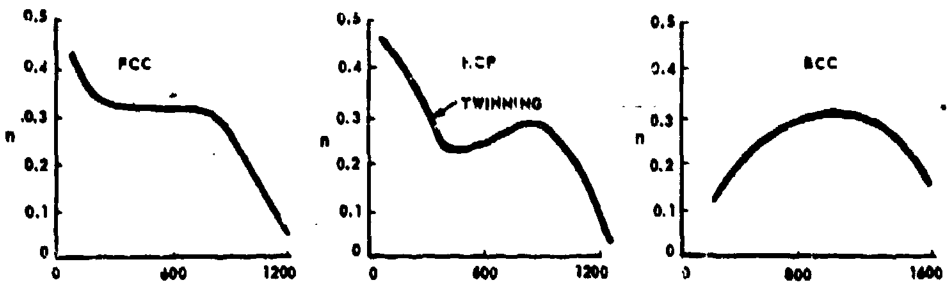


FIGURE A-37. TYPICAL CURVES OF STRAIN-HARDENING EXPONENT AS A FUNCTION OF TEMPERATURE FOR THREE SPACE LATTICES

Recrystallization

As cold-worked metals are heated to elevated temperatures, several processes occur, in the following order:

(1) Recovery, manifested by reduction of residual stresses resulting from a change in substructure

(2) Recrystallization, the principal softening process which reduces the hardness and strength, and increases the ductility

(3) Grain growth

(4) Secondary recrystallization, or creation of new grain-orientation textures by preferential grain growth.[26]

Recovery and recrystallization are believed to be intimately related to the process of polygonization, which is the formation of substructure by rearrangement of dislocations on bent lattice planes.

Recrystallization should be regarded as occurring over a temperature range, which is a function of degree of strain hardening, time of annealing at a temperature, and solute content in solid solution. The recrystallization temperature range can be crudely estimated as $T_m/2$, where T_m is the absolute melting point. This is also roughly the temperature at which the rate of self-diffusion becomes appreciable and recovery from strain hardening by dislocation climb becomes dominant, characteristics of high-temperature deformation. It is also roughly equal to the "equi-cohesive" temperature, which supposedly marks the change in strength of the grain relative to the grain boundary. Below this temperature, the grain boundaries are thought to be stronger than the grains and fracture is expected to occur in a transgranular fashion. Above this temperature, the grains are thought to be stronger than the grain boundaries and fracture is expected to occur in an intergranular mode, which is observed in practice.

Hot Working

Hot working is plastic deformation accomplished above the recrystallization range and is characterized by concurrent recovery of strain hardening and an essentially constant flow stress. Deformation is limited by initiation of ductile fracture or some form of hot shortness. This general response is essentially the behavior of an ideal plastic body. The flow stress, of course, decreases with increasing temperature and goes to zero abruptly at the melting point.

A characteristic of the hot-working range is the increased sensitivity of the strength to strain rate, or the property of "strain-rate sensitivity". This is illustrated for high-purity aluminum in Figure A-38.[27] It is of interest to note that the strengths at various temperatures approach a common value at very high strain rates.

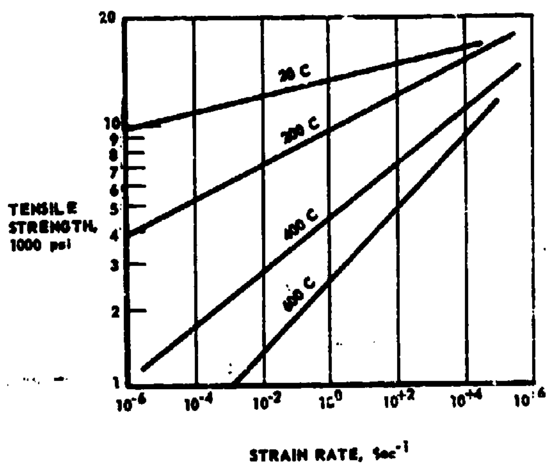


FIGURE A-38. RELATIONSHIP BETWEEN TENSILE STRENGTH AND STRAIN RATE FOR HIGH-PURITY ALUMINUM

The combined influences of temperature and strain rate on the relative strength of a pure metal or single-phase alloy is, in general, as shown in Figure A-39 for the homologous temperatures expressed as a fraction of the melting temperature, T_m, on an absolute scale. Up to 0.5 T_m the strain rate has little influence on strength properties. At about 0.5 T_m, recrystallization takes place. Between 0.5 T_m and T_m, the metals exhibit the greatest strain-rate sensitivity.

Typical influences of temperature and strain rate are shown in Figure A-40 for three single-phase metals representative of aluminum-base, titanium-base, and cobalt-base alloys. As shown, the behaviors are quite similar.

It is worth noting that alloys containing two phases exhibit slightly less predictable behavior with respect to temperature and strain rate. This is due primarily to the fact that one of the phases exhibits recrystallization temperatures different from the other. The alloys become increasingly strain-rate sensitive at temperatures above the lowest recrystallization temperature of the two phases rather than at the theoretical 0.5 T_m temperature. Steels and alpha-beta titanium alloys exhibit this behavior. Figure A-41 indicates how steels are generally influenced by strain rate and Figure A-42 illustrates the behavior of alpha-beta titanium alloys. At temperatures between absolute zero and about 900 F, the strengths of steels are virtually insensitive to strain-rate variation. Steels generally exhibit a hump on the curve at temperatures in the vicinity of the transformation temperature range. At higher rates this hump shifts towards higher temperatures, as shown in the figure. Titanium alloys exhibit a minor amount of strain-rate sensitivity at ambient temperatures. This is believed related to the metastable beta phase. Strain-rate sensitivity is evident at temperatures at about 0.45 T_m, and higher.

Work of Deformation

One of the "laws" of plastic deformation is that the sum of the three principal normal true strains is equal to zero, or

$$\phi_1 + \phi_2 + \phi_3 = 0 \quad . \qquad \text{(A-51)}$$

It follows that one of the strains must be of the opposite sign and equal to the negative sum of the other two strains. This strain parameter is known as the effective deformation, ϕ_h, and

$$\phi_h = -\Sigma\phi_n \quad . \qquad \text{(A-52)}$$

For simple geometries, the effective deformation is easily deduced. For example, in the compression of a solid cylinder, the effective deformation is the true strain in the axial direction and is negative in sign.

It can be demonstrated[28] that the work of deformation for uniaxial, biaxial, and triaxial stress states is given by the expression

$$A = V(K_f)\phi_h \quad , \qquad \text{(A-53)}$$

where

A is the work of deformation

V is the volume deformed

K_f is the flow stress

ϕ_h is the effective deformation.

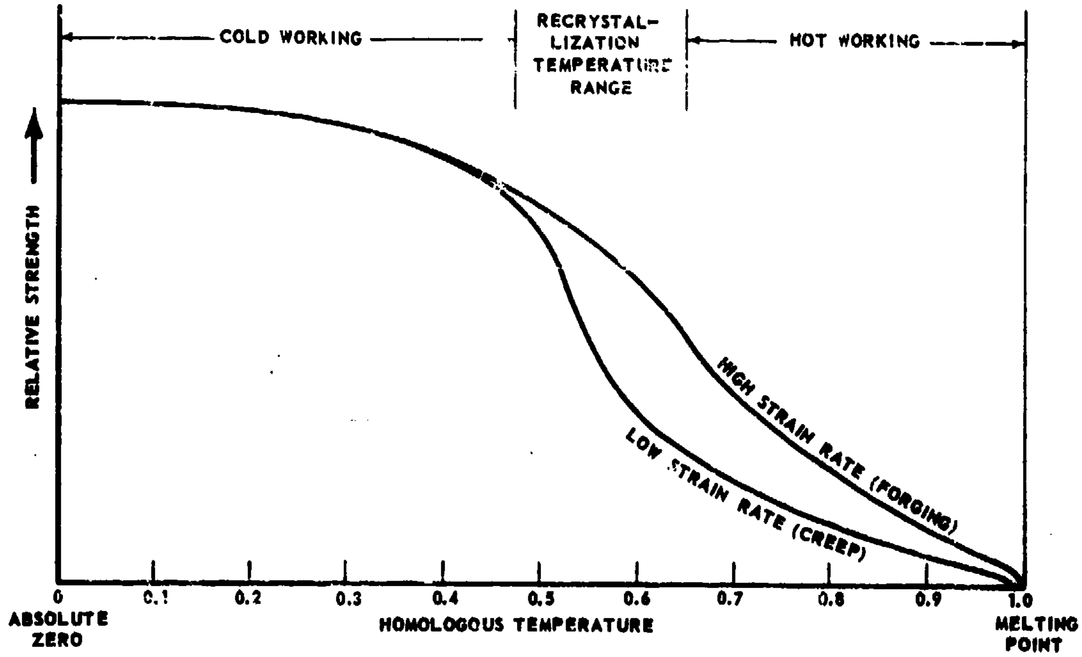


FIGURE A-39. TYPICAL INFLUENCE OF TEMPERATURE ON RELATIVE STRENGTH OF PURE METALS AND SINGLE-PHASE ALLOYS AT VARIOUS STRAIN RATES

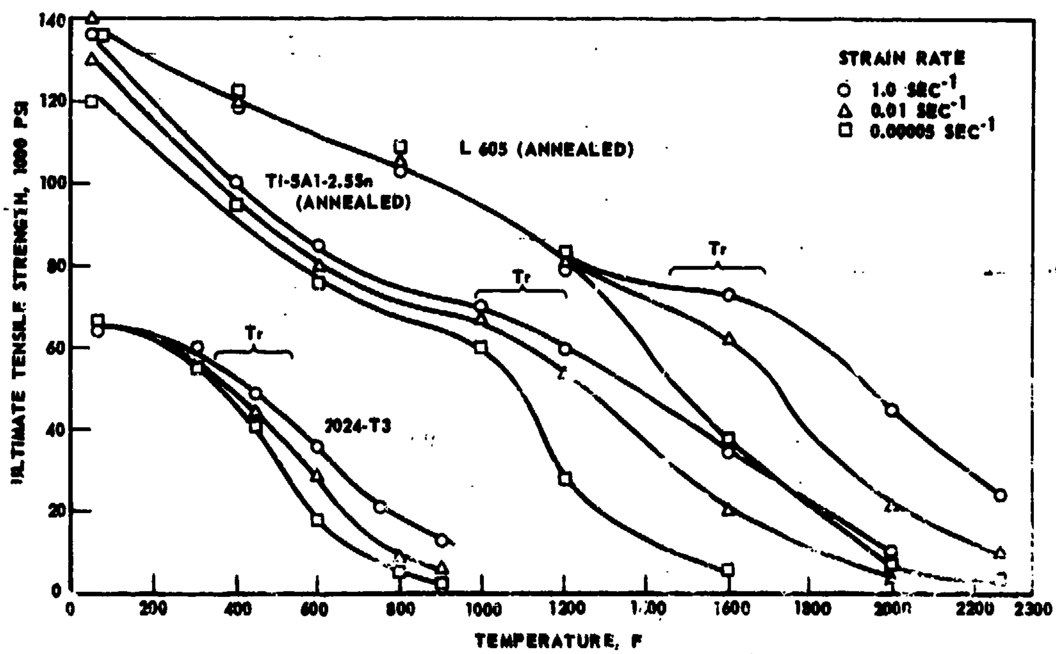


FIGURE A-40. TENSILE STRENGTHS OF TYPICAL ALUMINUM, COBALT, AND TITANIUM ALLOYS AS INFLUENCED BY TEMPERATURE AND STRAIN RATE

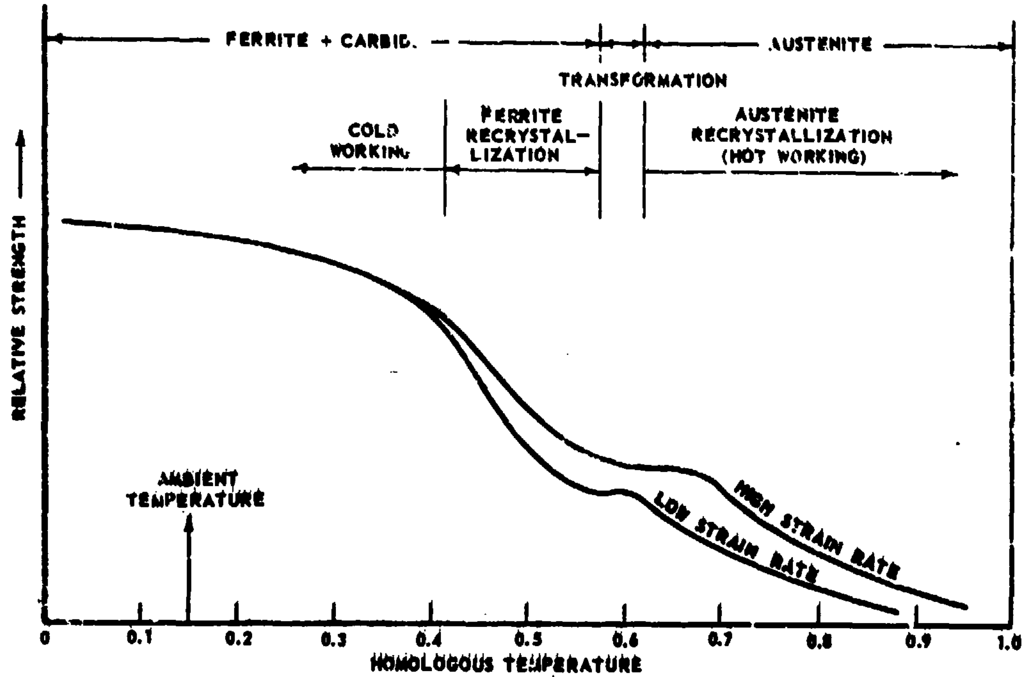


FIGURE A-41. INFLUENCE OF TEMPERATURE AND STRAIN RATE ON THE RELATIVE STRENGTH OF ANNEALED STEELS

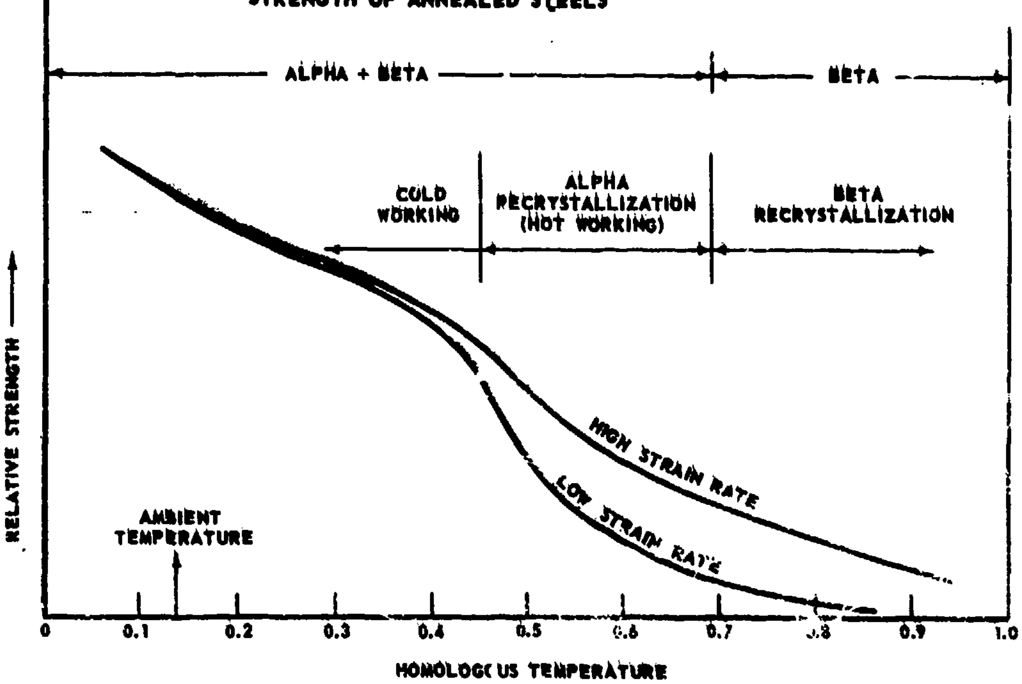


FIGURE A-42. INFLUENCE OF TEMPERATURE AND STRAIN RATE ON THE RELATIVE STRENGTH OF ANNEALED ALPHA-BETA TITANIUM ALLOYS

It was pointed out earlier that, for cold working, $K_f = f(\varphi_h)$, and for hot working, K_f = constant. This is shown graphically in Figure A-43(a). The specific work of deformation or the work per unit volume of material, and its dependence on strain for hot and cold working is depicted schematically in Figure A-43(b). The specific work varies linearly with strain in hot working, but increases curvilinearly in cold working.

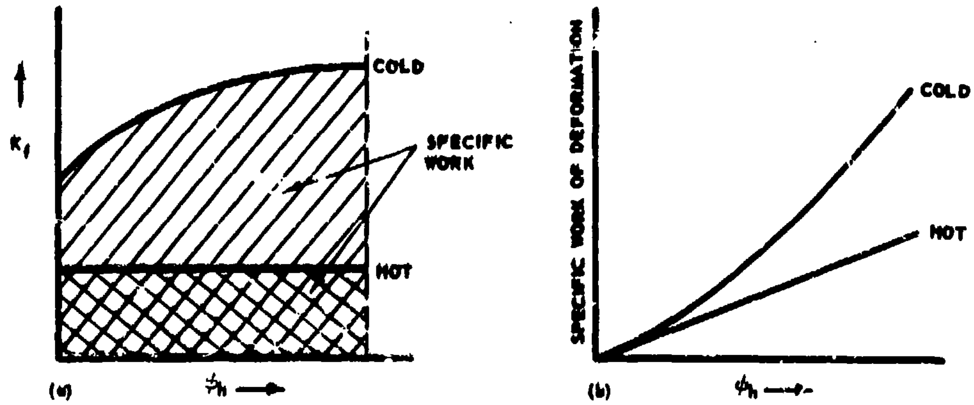


FIGURE A-43. THE SPECIFIC WORK OF DEFORMATION [CROSS-HATCHED AREA IN (a)] AND ITS DEPENDENCE ON STRAIN

Generation of Heat

The work of deformation is largely converted into heat. The temperature rise may be estimated[28] by the following relationship:

$$\Delta t = \frac{\phi_h \cdot K_f}{W \cdot c \cdot \gamma} , \qquad \text{(A-54)}$$

where

Δt is the temperature rise

$\phi_h \cdot K_f$ is the specific work of deformation

W is the mechanical equivalent of heat

c is the specific heat

γ is the density.

For iron at forging temperatures, $\Delta t \approx 36$ to 72 (ϕ_h). As the deformation temperature is decreased, the specific work of deformation increases, and Δt can be several times larger.

From a practical viewpoint, even with the sudden and severe deformations which occur in the rolling of wire rod or in forging of small sections under high-speed hammers, not only can an excessive temperature drop be prevented, but a temperature rise may be obtained, in spite of high radiation losses.

HARDENING MECHANISMS AND OTHER METALLURGICAL EFFECTS

Some metallurgical phenomena of particular importance in forging processes are discussed briefly in this section.

High-purity metals can be hardened by two mechanisms without the addition of alloying elements. These are, at a given temperature, (a) strain hardening and (b) reduction of grain size. The hardness, or strength, increases as the testing temperature decreases.

Strengthening by reduction of grain size results from the fact that, as a grain deforms by slip, it must adjust to its neighboring grains in order to maintain coherency; the smaller the grain size, the more the resistance to this adjustment in shape and the higher the resistance to flow. Hardness is generally related to the grain size in a nonlinear fashion but there is a linear relationship between hardness and the grain-boundary surface per unit volume. This illustrates the fact that increased hardness resulting from reduction of grain size is fundamentally related to grain boundaries.

Hardening by Alloying

The addition of alloying elements provides the basis for the following additional hardening mechanisms:

(1) Solid-solution hardening

(2) Second-phase hardening
  (a) Precipitation hardening
  (b) Dispersion hardening

(3) Strain aging

(4) Transformation hardening

(5) Ordering.

These mechanisms can have important influence on the forgeability of alloys, and plastic straining in turn can affect these hardening processes.

A solute atom may be completely soluble, partly soluble, or essentially insoluble in the lattice structure of the base metal. When the solute atom is present in amounts

less than the solubility limit, it will be found in solid solution, i.e., dissolved in the lattice of the host atom. When the solute atom is present in excess of its solubility, two structural components exist; namely, a saturated solid solution and a second phase composed of either the pure solute or a mixture of the solute and the solvent. The second phase may exist in a great variety of shapes, sizes, and distributions. The hardening of a metal by an alloying element, accordingly, occurs in two general ways:

(a) Formation of solid solutions, giving rise to solid-solution hardening, and

(b) Formation of a second phase, giving rise to hardening by interruption of the matrix by the second phase.

Solid-Solution Hardening. Two basic types of solid solutions are formed in metals, namely, "interstitial" solid solutions in which the solute atom enters into interstitial sites in the lattice, and "substitutional" solid solutions in which the host atom is replaced by the solute atom. Interstitial solid solutions in general follow Hagg's Rule, which states that the ratio of the radius of the solute atom to the radius of the solvent atom must be less than 0.59. These solid solutions are found to be largely restricted to C, O, N, H, and B in transition metals that have unfilled inner shells of electrons. Substitutional solid solutions in general follow, as one condition to be met, the Hume-Rothery Rule, which states that the ratio of the radius of the solute atom to the solvent atom must lie in the range of 0.85 to 1.15 (15 per cent rule).

Solid-solution hardening occurs fundamentally as a result of a decrease in the periodicity or regularity of the lattice. The effectiveness of hardening by a solute is found to be inversely proportional to its solubility. The solubility can be correlated with size effect, differences in electronegativity, and differences in valence. As the limit of solid solubility is approached, short-range ordering probably becomes important.

Second-Phase Hardening. When the solubility limit for a solute is exceeded, a second phase of the general composition AxBy appears in the microstructure, as illustrated in Figure A-44. The second-phase particles act as barriers to slip and thereby harden and strengthen the alloy. The second phase may come out of solution "coherently" or "noncoherently", depending on whether or not the lattice of the second phase is continuous or discontinuous with the matrix. When a noncoherent precipitate is formed, it is called a dispersed second phase and the process is called dispersion hardening. The strength of the alloy is inversely proportional to the mean free path in the matrix (between precipitated particles) when the second phase is harder than the matrix.

The hardening process is called "precipitation hardening" or "age hardening" when the precipitate forms coherently. The forced fit of the precipitate on the matrix lattice creates large strains over long distances and results in effective hardening. The particles grow with time and reach a critical size at which they cannot sustain the strains required; they then break away from the matrix and become noncoherent. A typical age-hardening system is illustrated in Figure A-45 and a typical aging curve in Figure A-46. Cold forging an age-hardenable alloy after solution quenching, but before aging, will accelerate aging, create favorable nucleation sites within the grain, and provide more general nucleation. With no deformation, the precipitation will nucleate at grain boundaries ahead of the sites in the interior of the grain, as illustrated in Figure A-47. Plastic deformation, thus, makes precipitation occur more uniformly.

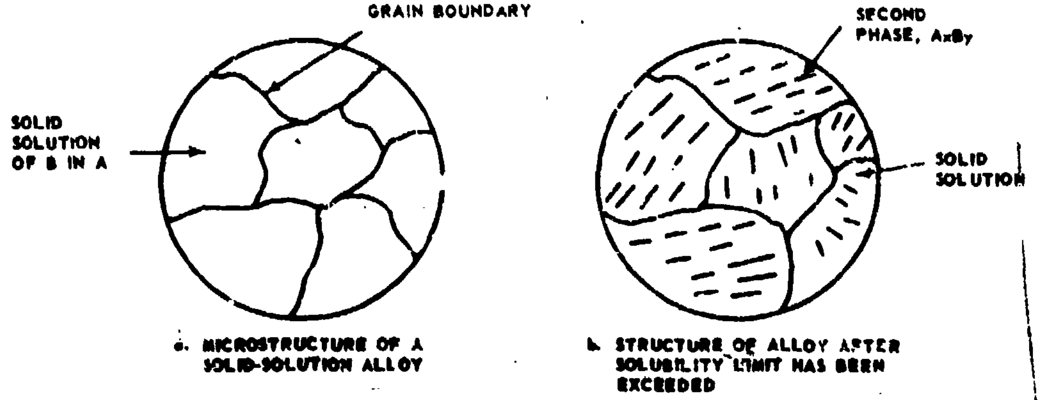


FIGURE A-44. CHANGE IN STRUCTURE ON EXCEEDING THE SOLID SOLUBILITY OF A SOLUTE

Strain Aging. The term "strain aging" is used in two senses. One refers to the return of the yield point after plastic straining of iron-base alloys. This is thought to be due to interstitial atoms locking dislocations by association with the dislocations. The other sense refers to the inducing of precipitation of, for example, iron nitride from iron, by cold working. The conditions of supersaturation are attained through cold working. Perhaps these two connotations differ only in the degree of completion of the same process.

Hardening by Allotropic Transformation. Several transition metals possess two or more crystalline modifications in the solid state, called allotropes. When one modification is cooled into the temperature range in which another allotrope is stable, a lattice transformation occurs that is called an allotropic transformation. In pure metals, such transformations effect only a small degree of hardening, as a result of grain refinement. In alloys, this is a very versatile hardening process because supersaturation in the low-temperature allotrope can be produced by cooling the high-temperature phase rapidly. This is true when the solubility of an element in the high-temperature allotrope is higher than in the low-temperature modification. If the quenched product is a supersaturated solid solution, it can be reheated to cause precipitation of a second phase either coherently or noncoherently. The system will relax to an equilibrium solid solution and a second phase. No new hardening mechanisms are involved; the lattice transformation serves only to achieve significant supersaturation. Plastic deformation is known to accelerate the decomposition of the high-temperature phase.

Order-Disorder Transformation. In a special type of transformation that occurs in some alloy systems, the lattice sites maintain the arrangement they have in a solid solution, but above a critical temperature the A and B atoms are randomly distributed, and below this temperature the atoms take up special sites with respect to their neighbors. This order-disorder transformation is illustrated in Figure A-45.(29) Ordering produces interesting electrical effects and is believed to be an important hardening mechanism in some complex alloys.

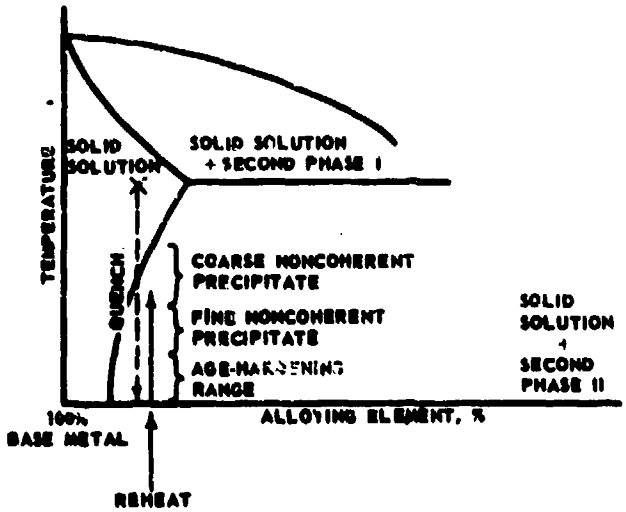


FIGURE A-45. STEPS IN THE AGE HARDENING OF AN ALLOY

Typical age-hardening system

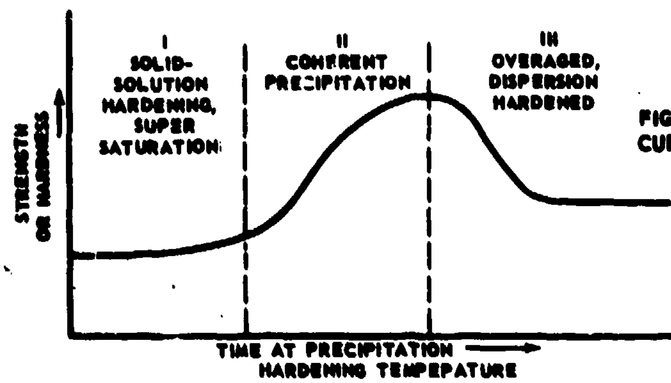


FIGURE A-46. TYPICAL AGING CURVE

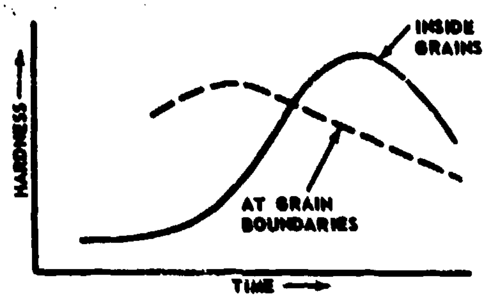


FIGURE A-47. SCHEMATIC ILLUSTRATION OF KINETICS OF AGING INSIDE GRAINS AND AT GRAIN BOUNDARIES

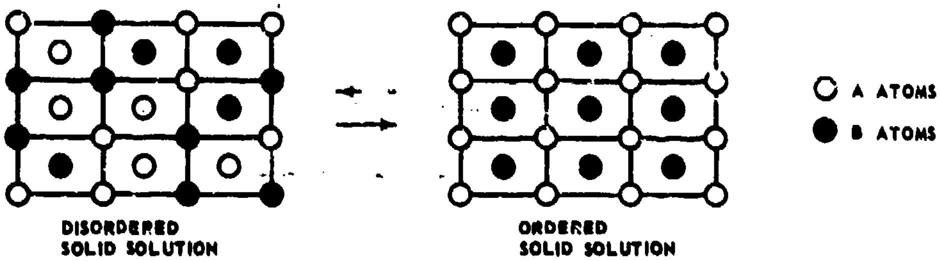


FIGURE A-48. SCHEMATIC ILLUSTRATION OF DISORDER-ORDER TRANSFORMATION

Stability of Microstructures at Elevated Temperatures

The stability of various elements of microstructure and hardening mechanisms at elevated temperatures is of interest both in determining the flow stress (and therefore the forging forces required) and the load-bearing abilities. Elevated temperatures bring about many changes in microstructure and hasten the attainment of the equilibrium structures. Some of the expected effects are listed below:

(1) The rates of diffusion increase

(2) Grain coarsening occurs

(3) Solubilities usually increase, tending to dissolve second-phase particles

(4) Aging and overaging are accelerated

(5) Second-phase particles tend to agglomerate

(6) The strength of the matrix continuously decreases with temperature, barring special metallurgical effects, and goes to zero at the melting point

(7) The reaction with chemically reactive atmospheres increases, leading to contamination or loss of constituents to the atmosphere

(8) Solute atoms can be absorbed at grain boundaries or in grains at elevated temperatures.

The stabilities of variously hardened aluminum and aluminum alloys as a function of temperature are summarized by Sherby[30] in Figure A-49. The figure shows that, with the exception of SAP (sintered aluminum powder), all the variously hardened alloys lose their strength abruptly at about one-half of the absolute melting point. The solid-solution-hardened and dispersion-hardened alloys lose their strength first, whereas the cold-worked and precipitation-hardened alloys lose their strength more slowly with

increasing temperature. At about three-quarters of the melting point, the strength of all the alloys becomes comparable to that of pure aluminum. The case of SAP is somewhat special since it is a dispersion of Al_2O_3 in aluminum. Aluminum has a very small solubility for oxygen and therefore Al_2O_3 has little tendency to dissolve, explaining the stability of this material.

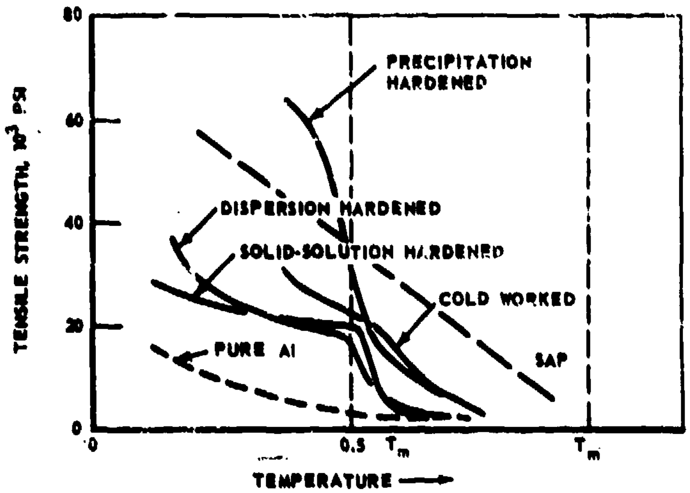


FIGURE A-49. EFFECT OF TEMPERATURE ON THE STABILITY OF VARIOUS HARDENING MECHANISMS

Figure A-49 illustrates strikingly the instability of various microstructures at elevated temperatures and the eventual disappearance of all the modes of hardening at sufficiently high temperatures.

Oxidation and Contamination

Since many elevated-temperature forging operations are performed in air, the workpiece is subject to oxidation, or scaling. This is a source of metal loss, but also can lead to surface defects and to important changes in friction coefficients.

The simplest type of oxidation is the formation of a continuous oxide film through which the metal ion diffuses to the oxide-atmosphere interface. The metal ionises at the metal-oxide interface and diffuses along with electrons to the oxide-atmosphere interface. The oxygen atom is ionised and reacts to form the oxide. In some refractory metals such as molybdenum, the oxygen ion diffuses through the oxide to the metal-oxide interface.

For a diffusion path X, the rate of oxidation is expected to be inversely proportional to X, or

$$dx/dt = K/x$$

and (A-55)

$$X^2 = Kt + \text{Constant} \quad .$$

This is the familiar "parabolic equation" for oxidation, and K is called the rate constant. The rate constant is expected to vary with temperature according to the Arrhenius relationship,

$$K = Ae^{-Q/RT} \quad , \tag{A-56}$$

where

A is a frequency factor

Q is an activation energy

R is the universal gas constant

T is the absolute temperature.

The rate constant varies exponentially with temperature, emphasizing the importance of temperature on the kinetics of oxidation. The volume of the scale in general is greater than the volume of metal from which it forms. At a critical thickness, the scale buckles and becomes loose. In the case of a continuous oxide film, when the temperature is too low to have significant ionic diffusion, the oxide film initially will form quickly and then slow down logarithmically to a limiting value:

$$X = K_1 \ln t + \text{Constant} \quad . \tag{A-57}$$

The limiting value for aluminum at room temperature is 40-100 Angstroms.

When there is appreciable solubility for oxygen and the alloy contains a species which is considerably more reactive than the base metal, a special type of contamination occurs, called "internal oxidation". The oxygen penetrates the alloy under the oxide and reacts with this solute, forming oxide particles in the alloy. These internal oxide particles can both strengthen and embrittle an alloy.

Of the chief sources of contamination of metals at elevated temperatures, oxygen, nitrogen, sulphur, and hydrogen, usually oxygen is most important. The dissociation constants of molecular hydrogen and nitrogen are very low, so that contamination from these elements is more likely if they are in a reactive form such as water vapor and ammonia, respectively. Carbon contamination can be important in atmospheres containing hydrocarbons. The product of reaction with the contaminant is generally a solid, giving rise to a second phase. Decarburization, or loss of carbon, is a special case in which the product of reaction is a gas. Alloying elements may be lost from the alloy if the species have significantly high vapor pressures.

Embrittlement of Metals and Alloys

The embrittlement of metals and alloys is a broad topic and the various phenomena can only be listed in this section. The term "embrittlement" generally implies a sharp drop in ductility with little change in other properties. The embrittling processes can be conveniently categorized into those resulting from (a) external constraints, (b) phase changes, and (c) gases.[31]

External Constraints

(1) Reduction of temperature, especially for body-centered cubic metals
(2) Increasing strain rate
(3) Imposition of triaxial stress state in loading
(4) Special environment, e.g., caustic embrittlement.

Phase Changes

(1) Strain aging and quench aging in low-carbon steel
(2) Precipitation, or solute concentration, at grain boundaries
(3) Blue brittleness, temper brittleness, and 500 F brittleness in steels
(4) Sigma-phase formation in stainless steels
(5) 885 F embrittlement of iron-chromium alloys.

Gases

(1) Embrittlement in solid solution
(2) Formation of porosity
(3) Formation of a second phase
(4) Formation of low-melting eutectics.

One of the important parameters that dictates the position of a solute atom in a structure and its potential effect on brittleness is its effect on grain-boundary energy. If a solute atom reduces the energy of a boundary, it prefers to be associated with the boundary, and if it increases the energy of the boundary, it prefers to be located away from the boundary. This is called positive and negative adsorption.(32) Those elements which are positively adsorbed to grain boundaries may show a positive or negative temperature coefficient of adsorption. If the temperature coefficient is positive, raising the temperature will force more solute to the grain boundaries. This behavior may impair ductility either through a solid-solution effect or the formation of a second phase on cooling down. Overheating of steels and perhaps temper brittleness in steels are examples of embrittlement through adsorption effects.

Residual Stresses

One of the important consequences of nonuniform plastic deformation of a metal is the generation of residual stresses, or "locked up" stresses in the section. These stresses are important because they represent an internal system of loading that can aid in failure by flow and fracture, and also cause distortion of the part during machining.

The formation of residual stresses in a cold-drawn bar formed by pulling it through a die is illustrated in Figure A-50. Usually the deformation is a little more severe in the outer zone, calling for this zone to contract when the external loading is removed. This zone develops residual tensile stresses because it is constrained from contracting by the core; the core develops residual compressive stresses. A condition of equilibrium demands that the total residual tensile load be equal to the total load in compression.

Residual stresses can also be developed by heating and cooling treatments which set up sufficiently high thermal gradients. The magnitude of residual stresses developed is limited by the flow stress at the temperature of deformation, since stresses in excess

of this value will relax by plastic flow. It follows, therefore, that the higher the temperature of deformation, the lower will be the magnitude of residual stresses developed. The same principle applies to stress-relief treatments. The higher the temperature used for a stress-relief treatment, the greater the decrease of residual stresses because of the reduction of the flow stress.

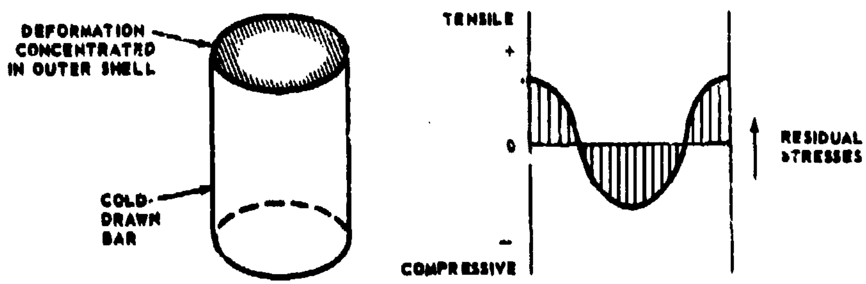


FIGURE A-50. FORMATION OF RESIDUAL STRESSES IN A COLD-DRAWN CYLINDRICAL BAR

ANISOTROPY IN WROUGHT ALLOYS

In contrast to the cast state, wrought metals can exhibit strong anisotropy of certain mechanical properties, particularly tensile ductility and toughness as measured by the impact test. It is important, therefore, to compare the structural features of the two states.

The important structural features are:

(1) Grain size
(2) Segregation
(3) Porosity
(4) Shape of inclusions.

Cast structures in general have coarser grain sizes, and grain refinement is expected to result from deformation. The segregation pattern is also on a coarser scale in the cast state. Segregation can be removed only by diffusion, which is very slow for substitutional solutes but reasonably rapid for interstitial elements. Plastic deformation has no particular effect on diffusion rates but will reduce the distance between regions of high and low alloy contents. This serves to reduce the diffusion distances and to increase concentration gradients, thus increasing the rate of homogenization.

Porosity is expected in cast structures unless extraordinary measures are taken to degas the liquid metal. Porosity may occur in more or less spherical blow holes or as fine-scaled "microporosity". The latter is the more important type in its effect on mechanical properties because it behaves as internal notches. Plastic deformation seals up these microporores and thus eliminates the internal-notch effect.

Nonmetallic inclusions are apt to be found as spherical or dihedral shapes in cast alloys, except for the special case of inclusions that come out as grain-boundary films. Deformation changes these inclusions into plates or stringers, which are ineffective stress raisers in the direction of working, but are effective in the direction transverse to the principal deformation direction.

The conversion of the cast state to the wrought state involves all four factors listed above. Porosity is quickly healed by deformation, as reflected by the rapid attainment of the maximum longitudinal value. The optimum transverse ductility is obtained at about a 2:1 forging reduction. Further deformation produces stringer inclusions and more pronounced and fibered segregation patterns. The orientation of segregation patterns in the transverse test coupons has an important effect on the type of fracture and the ductility values. The initial benefits of closing up porosity are diminished by the effects of stringer and segregation patterns.

REFERENCES - APPENDIX A

(1) Cottrell, A. H., Dislocations and Plastic Flow in Crystals, Oxford at the Clarendon Press (1956), p. 8 (after Frenkel).

(2) Freudenthal, A. M., The Inelastic Behavior of Engineering Materials and Structures, John Wiley and Sons, Inc., New York (1950), p. 74 (after H. Grimm and H. Wolff, R. Houwink).

(3) Petch, N. J., "The Fracture of Metals", Progress in Metal Physics, 5, Pergamon Press, Ltd., London (1954), p. 1 (after M. Polyani, E. Orowan, F. Seitz, and T. A. Read).

(4) Jaeger, J. C., Elasticity, Fracture, and Flow, Methven's Monographs on Physical Subjects, London (1956), p. ^9.

(5) See for example: (a) W. T. Read, Jr., Dislocations in Crystals, McGraw-Hill Book Co. (1953); (b) Dislocations and Mechanical Properties of Crystals, John Wiley and Sons, Inc. (1957), edited by J. C. Fisher, W. G. Johnston, R. Thomson, T. Vreeland, Jr.; (c) book by A. H. Cottrell, Reference 1.

(6) Taylor, G. I., "Plastic Strain in Metals", Journal of Institute of Metals, London, 62, 307 (1938).

(7) Barrett, C. S., Structure of Metals, Second Edition, McGraw-Hill Book Co., New York (1952), p. 337.

(8) Hall, E. O., "Twinning and Diffusionless Transformations in Metals", Butterworths Scientific Publications, London (1954), p. 42.

(9) McLean, D., Grain Boundaries in Metals, Monographs on the Physics and Chemistry of Materials, Oxford Press (1957), p. 81.

(10) Smith, C. S., "Metal Interfaces", American Society for Metals (1952), p. 88.

(11) Loc cit., p. 150.

(12) Sachs, G., Zur Ableitung einer Fliebedingung Zeit, Ver Dtsch. Ing., 72, 734 (1928).

(13) Kochendörfer, A., Plastische Eigenschaften von Kristallen und Metallischen Werkstoffen, Springer, Berlin (1941).

(14) Armstrong, R. W., "On Size Effects in Polycrystal Plasticity", J. Mech. Phys. Solids, 9, 196 (1961).

(15) Lankford, W. T., Low, J. R., and Gensamer, M., "The Plastic Flow of Aluminum Alloy Sheet Under Combined Loads", Metals Technology, AIME T.P. 2237 (August, 1947).

(16) Zener, C., and Hollomon, J. H., "Effect of Strain Rate Upon Plastic Flow of Steel", Journal of Applied Physics, 15, 22 (1944).

(17) Hollomon, J. H., "The Problem of Fracture", Welding Society Monographs, New York (1947).

(18) McAdam, D. J., and Mebs, R. W., "The Technical Cohesive Strength and Other Mechanical Properties of Metals at Low Temperatures", Proc. Amer. Soc. for Testing Materials, 43, 661-703 (1943).

(19) Reference 17, p. 29.

(20) Orowan, E., "Low Temperature Plasticity", Zeitschrift für Physik, 89, 605-659 (1934).

(21) Meacham, R. C., "Theories of Plastic Deformation of Single Crystals", Report A11-53N7 onr-358.

(22) Hollomon, J. H., and Zener, C., "Conditions of Fracture of Steel", Trans. Amer. Inst. of Min. & Met. Eng., 158, 283 (1944).

(23) Gensamer, Maxwell, Strength of Metals Under Combined Stresses, American Society for Metals, Cleveland, Ohio (1941).

(24) Pugh, S. F., "Relations Between the Elastic Moduli and the Plastic Properties of Polycrystalline Pure Metals", Philosophical Magazine, Ser. 7, 45 (367), 823 (August, 1954).

(25) Larson, F. R., and Kula, E. B., "Strain Hardening of High Purity Metals", Metallurgia, 64 (384), 159-164 (October, 1961).

(26) Dieter, G. E., Jr., Mechanical Metallurgy, McGraw-Hill Book Co., New York (1961), p. 153.

(27) Smith, G. V., Properties of Metals at Elevated Temperatures, McGraw-Hill Book Co., New York (1950), p. 88.

(28) Siebel, E., "The Plastic Forming of Metals", Part I, Steel (November 13, 1933), p. 26.

(29) Guy, A. G., Elements of Physical Metallurgy, Addison-Wesley Press, Inc., Cambridge, Mass. (1951), p. 94.

(30) Sherby, O. D., "Factors Affecting the High Temperature Strength of Polycrystalline Solids", Progress Report, Contract AF 33(616)-6789, Aeronautical Research Laboratory, Wright-Patterson Air Force Base, Ohio (December, 1961), p. 7.

(31) Queneau, B. R., The Embrittlement of Metals, American Society for Metals, Cleveland, Ohio (1956).

(32) See for example: D. McLean, Grain Boundaries in Metals, Oxford at the Clarendon Press (1957).

APPENDIX B

SELECTED BIBLIOGRAPHY ON FORGING PRACTICES

APPENDIX B

SELECTED BIBLIOGRAPHY ON FORGING PRACTICES

(1) Aluminum Company of America, Pittsburgh, Pennsylvania, "Designing for Alcoa Forgings" (1950).

(2) American Machinist, "Drafting Practice for Forgings", a series of Reference Book Sheets published from April 23, 1956, to July 30, 1956.

(3) American Society for Metals, Cleveland, Ohio, Metals Handbook, Eighth Edition, Volume 1, "Properties and Selections of Metals". Copyright 1961.

(4) Atroshenko, A. P., "Low-Waste Drop Forging of Round Forgings in Dies With a Wedge-Shaped Flash Groove", Kuznechno-Shtampovochnoye Proizvodstvo, 3 (6), 4-7 (1961). Battelle Translation.

(5) Battelle Memorial Institute, Columbus, Ohio, "A Study on Possible Methods for Improving Forging and Extruding Processes for Ferrous and Nonferrous Materials", Final Engineering Report for the United States Air Force on Contract AF 33(600)-26272 (June 30, 1957).

(6) Beaver, W. W., "Wrought Fabrication of Beryllium Metal", paper presented at Institute of Metals Conference at the Royal Commonwealth Society, London, England, October 16-18, 1961.

(7) Bereshkovskii, D. I., "New Method of Determining Forgeability of Metals", Brutcher Translation No. 4439 from Zavodskaya Laboratoriya, 24 (7), 858-860. Copyright 1958.

(8) Boulger, F. W., "Plastic Working", Mechanical Engineering, 83, 55-57 (June, 1961).

(9) Canal, J. R., and Kunkler, W. C., "Exotic Alloys Forged by Heavy Presses", Space Aeronautics, 35, 54-57 (April, 1960).

(10) Canal, J. R., and Kunkler, W. C., "Forgings in Missiles and Space Vehicles, Aircraft Production, 23 (5), 188-193 (May, 1961).

(11) Carnahan, D. R., and Visconti, J. A., "The Extrusion, Forging, Rolling, and Evaluation of Refractory Alloys", Final Report of Westinghouse Electric Corporation, Blairsville, Pennsylvania, for the United States Air Force on Contract AF 33(616)-8325 (October, 1962).

(12) "Cold Forged Superalloys", Steel, 139, 126 (December 3, 1956).

(13) Colton, R. M., "The Effects of Fabrication Procedure on Properties of Ti-6Al-4Sn- 0.5Fe-0.25Cu", Watertown Arsenal Laboratories, Watertown, Massachusetts; Lecture No. 8 presented at the Fifth Titanium Metallurgical Conference, New York University, New York, New York, September 14-15, 1959.

(14) Croan, L. S., and Rizzitano, F. J., "The Influence of Forging Temperature on Mechanical Properties of Al-V Titanium Alloys", Report WAL 401/268 (TA 2-055) for the Watertown Arsenal, Watertown, Massachusetts (February, 1957).

(15) Darby, P. F., "The Development of Optimum Manufacturing Methods for Columbium Alloy Forgings", Interim Reports of the Crucible Steel Company of America, Midland, Pennsylvania, for the United States Air Force on Contract AF 33(600)-39944 (February, 1962).

(16) Denny, J. P., and McKeogh, J. D., "Forging Unclad Beryllium", Journal of Metals, 13 (6), 432-433 (June, 1961).

(17) Denny, J. P., and Rubenstein, H. S., "The Forging of Jacketed and Bare Beryllium", Institute of Metals Paper No. 54, presented at the Conference on The Metallurgy of Beryllium, October 16-18, 1961.

(18) Department of Commerce, Office of Technical Services, Washington, D. C., "Metalworking Part III: Casting and Forging", OTS Selective Bibliography SB-462 (April, 1961).

(19) Erbin, E. F., and Broadwell, R. G., "Heat Treatment of Forgings and Heavy Bar Sections of the Ti-13V-11Cr-3Al All-Beta Alloy", Titanium Metals Corporation of America, Toronto, Ohio; paper presented at the Sixth Annual Titanium Metallurgical Conference, New York University, New York, New York, September, 1960.

(20) Everhart, J. L., "Hot Forged Parts", Material and Methods Manual No. 136, Material and Methods, 45 (3), 135-153 (March, 1957).

(21) Favre, A. E., "Designing Aluminum Forgings", Machine Design, 28 (2), 76-84 (January 26, 1956).

(22) Feldmann, H. D., Cold Forging of Steel, translation of German Edition by Hayward, A. M., published by Hutchinson and Company, Ltd., London, England (1961).

(23) Fetzer, M. C., and Broverman, I., "Methods for the Preparation of Forging Stock from Large Aluminum Alloy Ingots", Final Project Report No. MS PR 56-34 of the Kaiser Aluminum and Chemical Corporation, Oakland, California, for the United States Air Force on Contract AF 33(600)-23480 (September 7, 1956).

(24) Flannagan, B. J., "Forgings - Sure Path to Quality", Metalworking, 17 (8), 48-49 (August, 1961).

(25) Generson, I. G., "Effect of Final Forging Temperature on Mechanical Properties of (Turbine Disk) Forgings", Brutcher Translation No. 4322 from Metallovedenie i Obrabotka Metallov, No. 8 (August, 1958).

(26) Gouwens, P. R., "Heated Dies Produce Better Forgings", American Machinist, 103 (22), 128-131 (October, 1959).

(27) Gowen, Edward F., Jr., "Beryllium Fasteners", Final Technical Engineering Report AMC-TR-60-7-807 of the Standard Pressed Steel Company, Jenkintown, Pennsylvania, for the United States Air Force on Contract AF 33(600)-39728 (August, 1960).

(28) Griest, A. J., Young, A. P., and Frost, P. D., "Relation Between Beta Grain Size and Ductility of High Strength Alpha-Beta Titanium Alloy", Transactions AIME, 215 (5), 844-848 (October, 1959).

(29) Gulyaev, A. P., Rustem, S. L., Orekhov, G. N., and Alekseeva, G. P., "New Die Steels for the Drop Forging of Creep-Resisting Steels and Alloys", Brutcher Translation No. 4274 from Metallovedenie i Obrabotka Metallov, No. 7, 2-10 (1958).

(30) Gure, Charles, "Designing Forged Parts", Product Engineering, 33 (9), 47-52 (April 30, 1962).

(31) Hamer, J. E., "Investigation to Develop Optimum Properties in Forged Ti-7Al-4Mo", Report WADD-TR-60-489 from the Crucible Steel Company, Midland, Pennsylvania, for United States Air Force on Contract AF 33(616)-6122 (October, 1960).

(32) Hammarlund, P. E., and Sundberg, Y., "Induction Heating for Forging", (Sweden A.S.E.A. Journal 1957) Metal Treatment and Drop Forging, 25, 249-253 (June, 1958).

(33) Hayes, A. F., and Yoblin, J. A., "Beryllium Forging Program", Final Report ASD-TR-62-7-647 of the Ladish Company, Cudahy, Wisconsin, for the United States Air Force on Contract AF 33(600)-36795 (June, 1962).

(34) Hayes, A. F., and Yoblin, J. A., "New Concepts for Fabricating Beryllium Through Advanced Forging and Powder-Metallurgy Technique", paper presented at Institute of Metals Conference at the Royal Commonwealth Society, London, England, October 16-18, 1961.

(35) Heim, A. I., "Hot-Press Forging Cuts Cost of Copper Alloy Parts", Iron Age, 184 (4), 74-76 (July 23, 1959).

(36) Henning, H. J., and Boulger, F. W., "Superalloy Forgings", DMIC Memorandum 86, Battelle Memorial Institute, Columbus, Ohio (February 10, 1961); OTS PB 161236.

(37) Henning, H. J., and Boulger, F. W., "High-Strength-Steel Forgings", DMIC Report 143, Battelle Memorial Institute, Columbus, Ohio, to the Office of the Director of Defense Research and Engineering (January 5, 1961); OTS PB 151102.

(38) Henning, H. J., and Frost, P. D., "Titanium-Alloy Forgings", DMIC Report 141, Battelle Memorial Institute, Columbus, Ohio, to the Office of the Director of Defense Research and Engineering (December 19, 1960); OTS PB 151100.

(39) Henning, H. J., Sabroff, A. M., and Boulger, F. W., "Evaluation of Molybdenum Alloy Extrusions", report by Battelle Memorial Institute, Columbus, Ohio, on Subcontract with Harvey Alu inum, Inc., Torrance, California, on Contract AF 33(616)-6377 (May, 1961).

(40) Hicks, J. S., "Establishment of Production Methods for Aircraft Quality Precision Forgings in Titan. m Alloys", AMC Technical Report 59-7-559 of North American Aviation, Incorporated, Los Angeles, California, for the United States Air Force, Wright-Patterson Air Force Base on Contract AF 33(600)-33703 (November, 1959).

(41) "How to Forge Molybdenum", Steel, 143, 8[illegible] (November 3, 1958).

(42) "How to Get More for Your Forging Dollar", Iron Age, 182 (8), 89-104 (August 21, 1958).

(43) Jablonski, S. J., "Magnesium Forging Developments Provide Greater Economics and Wider Application Potential", paper presented at the Seventeenth Annual Convention of the Magnesium Association, New York, New York, October 16-18, 1961.

(44) Johnson, C. G., Forging Practice, published by American Technical Society, Chicago, Illinois. Copyright 1932.

(45) Lake, F. N., and Bresnyak, E. J., "Tungsten Forging Development Program", Third Interim Technical Progress Report ASD-TR-7-797 of Thompson Ramo Wooldridge, Inc., Cleveland, Ohio, for the United States Air Force on Contract AF 33(600)-41629 (March, 1961).

(46) Lippmann, H., and Neuberger, F., "Forces Effective in Dies", Werkstattstechnik u. Machinenbau, 48, 449-452, (August, 1958).

(47) Loria, Edward A., "Properties of Large Steel Forgings", Steel Processing and Conversion, 43 (4), 193-198 (April, 1957).

(48) Luchok, J., "How to Forge Astroloy", Metal Progress, 79 (2), 91-93 (February, 1951).

(49) Maurer, E., and Korshan, H., "Contribution to the Knowledge of the Mechanical Properties of Very Large Forgings", Stahl u. Eisen, 53, 209, 243, 271, (March 1953).

(50) McDaniel, J. L., "Pneumatic Mechanical Forming", paper presented at the ASTME Creative Manufacturing Seminar, New York University, New York, New York, 1960-1961.

(51) Moore, J. B., and Luchok, John, "Superalloy Forging Leap Size Barrier", Steel, 146, 72-73 (January 11, 1960).

(52) National Academy of Science, Washington, D. C. "State-of-the Art Report from Panel on Forging and Extrusion of the Committee of the Development of Manufacturing Processes for Aircraft Materials (AMC)", on Contract DA 36-039-sc-76436, (October 20, 1961). AD 248985.

(53) Naujoks, W., and Fabel, D. C., Forging Handbook, published by the American Society for Metals, Cleveland, Ohio. Copyright 1939.

(54) Nemy, A. S., and Grobe, A. H., "Tungsten Forging Development Program", Second Interim Technical Progress Report of Thompson Ramo Wooldridge, Inc., Cleveland, Ohio, for the United States Air Force on Contract AF 33(600)-41629 (December 12, 1960).

(55) "New Press Sweeps Hot Forging into the Automation Age", Iron Age, 182 (15), 82-83 (October 2, 1958).

(56) Nimmer, E. H., "Forging Program to Improve Mechanical Properties of Large Steel Forgings", Final Technical Engineering Report AMC-TR-60-7-755 of the Ladish Company, Cudahy, Wisconsin, for the United States Air Force on Contract AF 33(600)-38060 (December, 1960).

(57) Ogden, H. R., Maykuth, D. H., et al., "Scale-Up Development of Tantalum-Base Alloys", Final Report ASD-TR-61-684 of Battelle Memorial Institute, Columbus, Ohio, for the United States Air Force on Contract AF 33(616)-7452 (March, 1962).

(58) Palsulich, J. M., and Headman, M. L., "Developments in Pneumatic High Energy-Rate Forming", paper presented at the ASTME Creative Manufacturing Seminar, Detroit, Michigan, September 27-28, 1961.

(59) Peel, E. W., "Aluminum Forgings", Metal Industry, London, 86, 472-475 (June 3, 1955).

(60) Peel, E. W., "Close-to-Form and Close-Tolerance Forgings", Metal Treatment and Drop Forging, 28 (185), 53-58 (February, 1961).

(61) Petersen, A. H., and Pipher, F. C., "Forging Molybdenum by High-Energy-Rate Method", Metal Progress, 82 (2), 78-82 (August, 1962).

(62) Pflaume, E., "Timing for Forgings as a Function of Resistance and Reheating Time", Fertigungstechnik, 6, 541-543 (December, 1956).

(63) Pipher, F. C., "Molybdenum High-Rate Energy Precision Forging Development", Report LR 15290 of the Lockheed Aircraft Corporation, Burbank, California (July 31, 1961).

(64) "Precision Cold Forging - Super-Alloys Successfully Cold Forged", Steel Processing and Conversion, 43 (2), 77-78 (February, 1957).

(65) "Precision Forging", Metal Treatment and Drop Forging, 27 (180), 363-365 (September, 1960).

(66) "Precision Forging Up to Date", Metal Progress, 77 (5), 125-131 (May, 1960).

(67) Proffitt, C. E., and Thomas, H. C., "Close Tolerance Steel Forging Development", Final Technical Engineering Report ASD-TR-62-7-546 of the Boeing Company, Wichita, Kansas, for the United States Air Force on Contract AF 33(600)-36659 (April, 1962).

(68) "Reflectors Save Heat During Forging", Metal Treatment and Drop Forging, 28 (185), 58 (February, 1961).

(69) Rhoades, R. C., "New Moly Reference Blocks Improve Sonic Testing", Iron Age, 186 (17), 94-95 (October 27, 1960).

(70) Riordan, T. S., "High Energy-Rate Forming in Production at Bendix-Utica", paper presented to the ASTME Creative Manufacturing Seminar, Detroit, Michigan, September 27-28, 1961.

(71) Russ, J. J., "Forgers Tame Refractory Metals", Iron Age, 185 (20), 162-164 (May 19, 1960).

(72) Semchyshen, M., and Barr, R. Q., "Investigation of the Effect of Hot-Cold Work on the Properties of Molybdenum Alloys", Report WADC-TR-56-454 of the Climax Molybdenum Company of Michigan, Detroit, Michigan, for the United States Air Force on Contract AF 33(616)-2861 (January, 1957). AD 110708.

(73) Semchyshen, M., Barr, R. Q., and McArdle, G. D., "Effect of Thermal Mechanical Variables on the Properties of Molybdenum Alloys", Report WADC-TR-60-451 of the Climax Molybdenum Company of Michigan, Detroit, Michigan, for the United States Air Force on Contract AF 33(616)-5447 (November, 1957). AD 251721.

(74) Semchyshen, M., McArdle, G. D., and Barr, R. Q., "Development of High Strength and High Recrystallization Temperatures in Molybdenum-Base Alloys", Report WADC-TR-58-551 of the Climax Molybdenum Company of Michigan, Detroit, Michigan, for the United States Air Force on Contract AF 33(616)-2861 (February, 1959). AD 209383.

(75) Shofman, L. A., "On the Technology of Closed-Die Forging of Large Parts", Kuznechno Shtampovochnoye Proizvodstvo, 3 (6), 1-4 (1961).

(76) Showalter, H. L., "Precision Forging", Steel Processing and Conversion, 43 (5), 615-624 (November, 1957).

(77) Siebel, E., "The Present State of Scientific Knowledge in Hot Shaping or Forging", Steel Processing and Conversion, 43 (1), 19-24 (January, 1957).

(78) Simonson, R. B., Long, J. R., Tombaugh, R. M., and Green, R. C., "Development of Optimum Methods for the Primary Working of Refractory Metals", Final Engineering Report TR 60-418 of Harvey Aluminum Incorporated, Torrance, California, on United States Air Force Contract AF 33(616)-6377 (January, 1961).

(79) Singleton, R. H., Bolin, E. L., and Carl, F. W., "The Fabrication of Complicated Refractory Metal Shapes by Arc-Spray Techniques", Report by Allison Division of General Motors, Indianapolis, Indiana, paper presented at the High Temperature Materials Conference in Cleveland, Ohio, April 26-27, 1961.

(80) Smith, C. H., Jr., "The Forging Industry and the Future", Steel Processing and Conversion, 43 (10), 576-580, 635 (October, 1957).

(81) Spencer, L. F., "Selection of Materials for Hot Work Application, Part IV", Steel Processing and Conversion, 43 (8), 447-469 (August, 1957).

(82) Spencer, L. F., "Tool Steels, Basic Concepts Governing Their Treatment, Part I", Steel Processing and Conversion, 43 (5), 253-259 (May, 1957).

(83) Sprague, L. E., "The Effect of Vacuum Melting on the Fabrication and the Mechanical Properties of Forgings", paper presented at the Symposium on Vacuum Metallurgy at New York University, New York, New York (June 2, 1960).

(84) Stone, M. D., "Design and Construction of Large Forgings and Extrusion Presses for Light Metals", Transactions ASME, 75, 1439 (1953).

(85) "90-10 Tantalum-Tungsten Alloy Forged in Air", Materials in Design Engineering, 51 (2), 11 (February, 1960).

(86) Tikhomirov, N. V., "Forging Practice and Quality of Steel for Steam-Turbine and Turbogenerator Rotors", Brutcher Translation No. 4186 from Metallovedenie i Obrakotka Metallov, No. 4, 39-43 (April, 1958).

(87) Trabold, A. F., and Bank, S., "Fabrication Studies of Columbium Alloy Sheet", Metal Progress, 79 (5), 103-107 (May, 1961).

(88) "Vacuum Pour Giant Ingots to Make Flake-Free Forgings", Iron Age, 182 (18), 81-83 (October 30, 1958).

(89) Voss, H., "Relations Between Primary Structure, Reduction in Forging, and Mechanical Properties of Two Structural Steels", Arch. Eisenhüttenw., 7, 403-406 (January, 1934).

(90) Waine, A. H., and Rait, J. R., "The Forging of Heat Resisting Metals", Metal Treatment and Drop Forging, 25 (152), 191-196 (May, 1958).

(91) Watmough, T., Berry, J. T., and Gouwens, P. R., "Improvement of Mechanical Properties of Steel Castings by Press Forging", Final Technical Report AMC-TR-60-7-637 of the Armour Research Foundation for the United States Air Force on Contract AF 33(600)-36387 (September, 1960).

APPENDIX C

SELECTED BIBLIOGRAPHY ON PLASTIC DEFORMATION AND METAL-FLOW THEORY

APPENDIX C

SELECTED BIBLIOGRAPHY ON PLASTIC DEFORMATION AND METAL FLOW THEORY

(1) Alder, J. F., and Phillips, V. A., "The Effect of Strain Rate and Temperature on the Resistance of Aluminum, Copper, and Steel to Compression", Journal of the Institute of Metals, 83, 80-86 (1954-1955).

(2) Anderson, C. T., Kimball, R. W., and Cattoir, F. R., "Effect of Various Elements on Hot-Working Characteristics and Physical Properties of Fe-C Alloys", Journal of Metals, 5 (4), 525-529 (April, 1953).

(3) Andrade, E. N., "The Physics of the Deformation of Metals", Endeavour, 8-9 (36), 165-177 (October, 1950).

(4) Andreev, I. A., Barusdin, and Gluskin, L. Ia., "On the Estimation of the Plasticity of Low-Carbon Alloy Steel by the Hot Upsetting Method", Zavodskaya Laboratoriya, 24, 855 (July, 1958).

(5) Armstrong, R. W., "On Size Effects in Polycrystal Plasticity", Journal of the Mechanics and Physics of Solids, 9 (3), 196 (1961).

(6) Barton, J. W., Bodsworth, C., and Halling, J., "The Use of Paraffin Wax as a Mold Material to Stimulate the Plastic Deformation of Metals", Journal of the Iron and Steel Institute, 188, 321-330 (April, 1958).

(7) Beresovski, D. I., "Speed of Deformation of Forging and Hot Stamping", Vestnik Mashinostoeniya, 37, 63-66 (March, 1957).

(8) Boas, W., An Introduction to the Physics of Metals and Alloys, John Wiley and Sons, Inc., New York, New York. Copyright 1947; p 71

(9) Butler, L. H., "The Effect of Interposed Lubricants on the Surface Deformation of Metals During Plastic Working", Journal of the Institute of Metals, 88, 337-343 (April, 1960).

(10) Clark, C. L., and Russ, J. J., "A Laboratory Evaluation of the Hot-Working Characteristics of Metals", Transactions AIME, 167, 736-749 (1946).

(11) Conrad, Hans, "Effect of Temperature on Yield and Flow Stress of B.C.C. Metals", Philosophical Magazine, 5, 745-751 (July, 1960).

(12) Contractor, G. P., and Morgan, W. A., "Forgeability of Steels: A Critical Survey of the Literature", Metal Treatment and Drop Forging, 26 (161), 65-71 (February, 1959).

(13) Cook, M., and Larke, E., "Resistance of Copper and Copper Alloys to Homogeneous Deformation in Compression", Journal of the Institute of Metals, 71, 371 (1945).

(14) Cottrell, A. H., "Deformation of Solids at High Rates of Strain", Chartered Mechanical Engineering, 4 (11), 448-460 (November, 1957).

(15) Cottrell, A. H., Dislocations and Plastics Flow in Crystals, The International Series of Monographs on Physics, Clarendon Press, Oxford, England. Copyright 1953.

(16) Dieter, G. E., Jr., Mechanical Metallurgy, McGraw-Hill Book Company, Inc., New York, New York. Copyright 1961; p 153.

(17) Drapor, A. B., "Evaluation of Tests for Forgeability", Metal Treatment and Drop Forging, 24 (147), 495-502 (December, 1957).

(18) Ellis, O. W., "An Investigation into the Effect of Constitution on the Malleability of Steel at High Temperatures", Journal of the Iron and Steel Institute (Carnegie Scholarship Memoirs), 13, 47-81 (1924).

(19) Florentino, R. J., Sabroff, A. M., and Frost, P. D., "Development of Improved Methods for Cold Extrusion of Titanium", Final Engineering Report AMC-TR-61-7-557 of Battelle Memorial Institute, Columbus, Ohio, for the United States Air Force on Contract AF 33(600)-33540 (March, 1961).

(20) Freudenthal, A. M., The Inelastic Behavior of Engineering Materials and Structures, John Wiley and Sons, Inc., New York, New York. Copyright 1950; p. 74.

(21) Gensamer, Maxwell, "Strength of Metals Under Combined Stresses", paper presented at the ASM Twenty-Second National Metal Congress and Exposition, Cleveland, Ohio, October 21-25, 1940.

(22) Gubkin, S. I., "Similarity in Conditions of Deformation in the Working of Metals by Pressure", Brutcher Translation No. 2033 from Bulletin de L'Academie Des Sciences de L'ures, Classe de Sci. Tech., No. 1, 117-123 (1947).

(23) Gubkin, S. I. (editor), Theory and Practice of the Pressworking of Metals, Translation from Russian prepared by Technical Documents Liaison Office, MCLTD, Wright-Patterson Air Force Base, Ohio. Copyright 1956.

(24) Guy, A. H., Elements of Physical Metallurgy, Addison-Wesley Press, Inc., Cambridge, Massachusetts. Copyright 1951; p 44.

(25) Hall, E. O., Twinning and Diffusionless Transformations in Metals, Butterworths Scientific Publications, London, England. Copyright 1954.

(26) Hill, R., The Mathematical Theory of Plasticity, Clarendon Press, Oxford, England. Copyright 1950; pp 277-278.

(27) Hoffman, O., and Sachs, G., Introduction to Theory of Plasticity for Engineers, McGraw-Hill Book Company, Inc., New York, New York. Copyright 1953.

(28) Hollomon, J. H., The Problem of Fracture, American Welding Society, New York, New York. Copyright 1946.

(29) Hollomon, J. H., and Zener, C., "Conditions of Fracture of Steel", Transactions AIME, 158, 283 (1944).

(30) Ihrig, H. K., "A Quantitative Hot Workability Test for Metals", Iron Age, 153 (16), 86-89, 170 (April, 1944).

(31) Ihrig, H. K., "The Effect of Various Elements on the Hot Workability of Steel", Transactions AIME, 167, 749-777 (1946).

(32) Jaeger, J. C., Elasticity, Fracture, and Flow Methuen and Co. Ltd., London, England; John Wiley and Sons, Inc., New York, New York. Copyright 1956; pp 99-106.

(33) Johnson, W., "Indentation and Forging and the Action of Nasmyth's Anvil", Engineer, 205 (5328), 348-350 (March 7, 1958).

(34) Johnson, W., and Kudo, H., "The Compression of Rigid-Perfectly-Plastic Material Between Rough Parrallel Dies of Unequal Width", International Journal of Mechanical Sciences, 1 (4), 336-341 (July, 1960).

(35) Jordan, T. F., and Thomsen, E. G., "Comparison of an Unsymmetric Slip-Line Solution in Extrusion with Experiment", Journal of the Mechanics and Physics of Solids, 4 (3), 184-190 (1956).

(36) Kent, W. L., "The Behavior of Metals and Alloys During Hot Forging", Transactions AIME, 34, 209-226 (1928).

(37) Kienzle, Otto, "German Research in Drop Forging", Metal Treatment and Drop Forging, 20 (95), 361-368 (August, 1953).

(38) Knauschner, A., "Influence of Die Temperature and Die Lubrication on the Rising of Steel in the Die", Fertigungstech. u. Betrieb, 10, 139-145 (March, 1960).

(39) Kobayashi, S., Herzog, R., Lapsley, J. T., and Thomsen, E. G., "Theory and Experiment of Press Forging Axisymmetric Parts of Aluminum and Lead", Journal of Engineering for Industry, Series B, 81 (3), 228-238 (August, 1959).

(40) Kochendörfer, A., Plastische Eigenschaften von Kristallen und Metallische Werkstoffer, Springer Verlag, Berlin, Germany. Copyright 1941.

(41) Lankford, W. T., Low J. R., and Gensamer, M., "The Plastic Flow of Aluminum Alloy Sheet Under Combined Loads", Technical Publication No. 2237 (August, 1947), paper presented at a meeting of AIME, November, 1946.

(42) Larsen, F. R., and Kula, E. B., "Strain Hardening of High Purity Metals", Metallurgica, 64 (384), 159-164 (October, 1961).

(43) Lee, E. H., "The Theoretical Analysis of Metal Forming Problems in Plane Strain", Paper No. 51-A-8 reviewed by the ASME (November 29, 1950).

(44) Manjoine, M. J., "Influence of Rate of Strain and Temperature on Yield Stresses of Mild Steel", Transactions AIME, 66, A-211 (1948).

(45) Martin, L. H., and Bieber, L. O., "Methods of Evaluating Hot Malleability of Nickel and High Nickel Alloys", Non-Ferrous Rolling Practice, AIME (1948).

(46) McDonald, A. G., Kobayashi, S., and Thomsen, E. G., "Some Problems of Press Forging Lead and Aluminum", ASME Transactions, Series B, 246-252 (August, 1960).

(47) McLean, D., Grain Boundaries in Metals, Monographs on the Physics and Chemistry of Materials, Clarendon Press, Oxford, England. Copyright 1957.

(48) Meacham, R. C., "Single Crystal Plasticity; a Survey: Part III, Theories of Plastic Deformation of Single Crystals", Report ALL-S3 (T.01, NR-041-032) of Brown University, Providence, R. I., for the United States Navy on Contract N7-onr-358, T.01, NR-041-032 (August 15, 1948).

(49) Nadai, A., and Manjoine, M. J., "High Speed Tension Tests at Elevated Temperatures, Parts II and III", Transactions ASME, 63 A71-A91 (1941).

(50) News Brief from Carpenter Steel Company, Reading, Pennsylvania, "Tensile-Impact Test Reveals Best Hot Working Temperature", Metal Progress, 78 (4), 65 (October, 1960).

(51) Orowan, E., "Low Temperature Plasticity", Zeitschrift für Physik, 89, 605-659 (1934).

(52) Petch, N. J., "The Fracture of Metals", Progress in Metal Physics, Fergamon Press Ltd., London, England; Interscience Publishers, Inc., New York, New York (1954) Vol 5, pp 1-52.

(53) Polakowski, N. H., "The Compression Test in Relation to Cold-Rolling", Journal of the Iron and Steel Institute , 163, 250-276 (November, 1949).

(54) Prager, W., and Hodge, P. G., Theory of Perfectly Plastic Solids, John Wiley and Sons, Inc., New York, New York. Copyright 1950.

(55) Pugh, H. L. D., and Watkins, M. T., "Some Strain-Rate Effects in Drop Forging Tests", paper presented at the Conference on the Properties of Materials at High Rates of [illegible]in, Institute of Mechanical Engineers, London, England, May, 1957.

(56) Pugh, S. F., "Relations Between the Elastic Moduli and the Plastic Properties of Polycrystalline Pure Metals", Philosophical Magazine, Series 7, 45 (367), 823-842 (August, 1954).

(57) Queneau, B. R., The Embrittlement of Metals, American Society of Metals, Cleveland, Ohio. Copyright 1956.

(58) Raskind, V. L., "Compression and Geometry of Deformation", Vestnik Machinostroeniya, 37, 55-58 (January, 1957).

(59) Rastegaev, M. V., "Ductility of High Temperature Strength Alloys as Affected by Structure and Stress State", Brutcher Translation No. 4280 from Metallovendenie i Obrabotka Metallov, No. 7, 30-35 (1958).

(60) Rossand, Claude, "Contribution to the Study of Hot Plastic Deformation of Steels", Metaux (Corrosion-Industries), 35 (415), 102-115 (March, 1960).

(61) Sachs, G., "Zur Ableitung eine Fliebbedingung", Zeitschrift des Vereines deutscher Ingonieure, 72 (22), 734-736 (June 2, 1928).

(62) Schroeder, W., and Webster, D. A., "Press-Forging Thin Sections: Effect of Friction, Area, and Thickness on Pressure Required", Transactions ASME, 71, 289-294 (1949).

(63) Sheotakov, S. N., and Karnow, M. Y., "Structure and Properties of Alloys After Vibrational Working", Henry Brutcher Translation No. 4281 from Metallovedenie i Obrabotka Metallov, No. 7 (July, 1958).

(64) Sherby, O. D., Stanford University, Palo Alto, California, Preliminary information on an United States Air Force Contract.

(65) Shield, R. T., "Plastic Flow in a Converging Conical Channel", Journal of Mechanics and Physics of Solids, 3 (3), 246-258 (1955).

(66) Siebel, E., and Hitchcock, J. H., "The Plastic Forming of Metals - Part I - Laws and Fundamentals of Plastic Forming; Part II - Forces and Flow of Material in Technical Forming Processes; Part III - Investigation of Distribution of Deformation and Stress in Technical Forming Processes", Steel (October 16, 1933, through May 7, 1934), a weekly series.

(67) Siebel, Erich, and Pomp, A., "Die Ermittlung der Formänderungsfestigkeit von Metallen durch den Stauchversuch", Mitteilunger aus den Kaiser-Wilhelm Institut für Eisenforschung, 9, 157 (1927).

(68) Smith, C. S., "Metal Interfaces", paper presented at the ASM Thirty-Third National Congress and Exposition, Detroit, Michigan, October 13-19, 1951.

(69) Smith, G. V., Properties of Metals at Elevated Temperatures, McGraw-Hill Book Company, Inc., New York, New York. Copyright 1950; p 88.

(70) Tardif, H. P., "Deformation Studies in Metalworking Processes", Steel Processing and Conversion, 43 (11), 626-632 (November, 1957).

(71) Taylor, G. I., "Plastic Strain in Metals", Journal of the Institute of Metals, 62 (1), 307-324 (1938).

(72) Terlechi, E., "Factors Influencing the Behavior of Certain Grades of Austenitic and Ferritic Steels on Hot, Plastic Deformation", Hutnik, 24, 222-2[illegible]7 (June, 19[illegible]).

(73) Tholander, E., "Forging Research in Sweden", Metal Treatment and Drop Forging, 28 (186), 89-94 (March, 1961).

(74) Thomsen, E. G., and Kobayashi, S., "Approximate Solutions to a Problem of Press Forging", Journal of Engineering Industry, 81 (3), 217-227 (August, 1959).

(75) Tomlinson, A., and Stringer, J. D., "The Closing of Internal Cavities in Forging by Upsetting", Journal of the Iron and Steel Institute, 188, 209-217 (March, 1958).

(76) Tomlinson, A., and Stringer, J. D., "Spread and Elongation in Flat Tool Forging", Journal of the Iron and Steel Institute, 193, 157-162 (October, 1959).

(77) Wistreich, J. G., and Shutt, A., "Theoretical Analysis of Bloom and Billet Forging", Journal of the Iron and Steel Institute, 193, 163 (October, 1959).

(78) Zener, C., and Hollomon, J. H., "Effect of Strain Rate Upon Plastic Flow of Steels", Journal of Applied Physics, 15 (1), 22-32 (January, 1944).

(79) Zernow, L., "Plastic Behavior of Polycrystalline Metals at Very High Strain Rates", Physics Review, 91, 233 (1953).

APPENDIX D

SELECTED BIBLIOGRAPHY ON ELEVATED-TEMPERATURE PROPERTIES OF METALS AND ALLOYS

APPENDIX D

SELECTED BIBLIOGRAPHY ON ELEVATED-TEMPERATURE PROPERTIES OF METALS AND ALLOYS

(1) Achbach, W. P., and Favor, R. J., "Design Information on Titanium Alloys for Aircraft and Missiles", DMIC Report No. 45, Battelle Memorial Institute, Columbus, Ohio (January 10, 1961).

(2) Alder, J. F., and Phillips, V. A., "The Effects of Strain Rate and Temperature on the Resistance of Aluminum, Copper, and Steel to Compression", Journal of the Institute of Metals, 83, 80-86 (1954-1955).

(3) Bernett, E. C., "Short Time, Elevated Temperature, Stress-Strain Behavior of Tensile, Compressive and Column Members", WADC-TR-59-484 of The Marquardt Corporation, Van Nuys, California, for the United States Air Force on Contract AF 33(616)-6043 (December, 1959). AD 232963.

(4) Campbell, J. E., Goodwin, H. B., Wagner, H. J., Douglas, R. W., and Allen, B. C., "Introduction to Metals for Elevated-Temperature Use", DMIC Report No. 160, Battelle Memorial Institute, Columbus, Ohio (October 27, 1961).

(5) Chang, W. H., "A Study of the Influence of Heat Treatment on Microstructure and Properties of Refractory Alloys", Final Technical Report ASD-TDR-62-211 of the General Electric Company, Cincinnati, Ohio, for the United States Air Force on Contract AF 33(616)-7125 (November, 1961). AD 278561.

(6) Clark, D. S., "Influence of Impact Velocity on the Tensile Characteristics of Some Aircraft Metals and Alloys", California Institute of Technology, Technical Notes NACA, No. 868 (October, 1942).

(7) Conrad, Hans, "Effect of Temperature on Yield and Flow Stress of B.C.C. Metals", Philosophical Magazine, Series 8, 5 (55), 745-751 (July, 1960).

(8) Cottrell, A. H., "Deformation of Solids at High Rates of Strain", Chartered Mechanical Engineering, 4 (11), 448-460 (November, 1959).

(9) Craighead, C. M., Grube, K. R., Eastwood, L. W., and Lorig, C. H., "The Effects of Temperature on the Mechanical Properties of Magnesium Alloys", A Technical Survey G-1112-11 by Battelle Memorial Institute, Columbus, Ohio, to Project RAND (April 29, 1949).

(10) "Data Sheet-Refractory Metals", Missile Design and Development, 7 (2), 65 (February, 1960).

(11) Dow Chemical Company, Midland, Michigan, Dow Metal Handbook. Copyright 1944.

(12) Dukes, W. H., Anthony, F. M., and Pearl, H. A., "Investigation of Feasibility of Utilizing Available Heat Resistant Material for Hypersonic Leading Edge Applications", Report No. 7008-0252-012 of the Bell Aircraft Corporation, Buffalo, New York, for the United States Air Force on Contract AF 33(616)-6034 (October 15, 1959).

(13) Favor, R. J., and Achbach, W. P., "Design Information on 5Cr-Mo-V Alloy Steels (H-11 and 5Cr-Mo-V Aircraft Steel) for Aircraft and Missiles", DMIC Report No. 116R, Battelle Memorial Institute, Columbus, Ohio (September 30, 1960).

(14) Favor, R. J., Roberts, D. A., and Achbach, W. P., "Design Information on Nickel-Base Alloys for Aircraft and Missiles", DMIC Report No. 132, Battelle Memorial Institute, Columbus, Ohio (July 20, 1960).

(15) Favor, R. J., Deel, O. L., and Achbach, W. P., "Design Information on PH 15-7Mo Stainless Steel for Aircraft and Missiles", DMIC Report No. 135, Battelle Memorial Institute, Columbus, Ohio (August 22, 1960).

(16) Favor, R. J., Deel, O. L., and Achbach, W. P., "Design Information on PH 17-7Mo Stainless Steel for Aircraft and Missiles", DMIC Report No. 137, Battelle Memorial Institute, Columbus, Ohio (September 23, 1960).

(17) Favor, R. J., Deel, O. L., and Achbach, W. P., "Design Information on AM-350 Stainless Steel for Aircraft and Missiles", DMIC Report No. 156, Battelle Memorial Institute, Columbus, Ohio (July 26, 1961).

(18) Fisher, G. H., Gideon, D. N., McClure, G. M., and Holden, F. C., "Development of Methods and Instruments for Mechanical Evaluation of Refractory Materials at High Temperature", Eighth Quarterly Progress Report of Battelle Memorial Institute, Columbus, Ohio, for the United States Air Force on Contract AF 33(616)-6155 (February 28, 1961).

(19) Griest, A. J., Sabroff, A. M., and Frost, P. D., "Effect of Strain Rate and Temperature on the Compressive Flow Stresses of Three Titanium Alloys", Transactions ASM, 51, 935-945 (1959).

(20) Hall, R. W., and Sikora, P. F., "Tensile Properties of Molybdenum and Tungsten From 2500 F to 3700 F", NASA Memorandum Memo 3-9-59E of the Lewis Research Center, Cleveland, Ohio, for the National Aeronautics and Space Administration (February, 1959). AD 213438.

(21) Hucek, H. J., Elsea, A. R., and Hall, A. M., "Evolution of Ultrahigh-Strength, Hardenable Steels for Solid-Propellant Rocket-Motor Cases", DMIC Report No. 154, Battelle Memorial Institute, Columbus, Ohio (May 25, 1961).

(22) Kent, W. L., "The Behavior of Metals and Alloys During Hot Forging", Transactions AIME, 34 (1928).

(23) Langston, M. E., and Lund, C. H., "Physical Properties of Some Nickel-Base Alloys", DMIC Report No. 129, Battelle Memorial Institute, Columbus, Ohio (May 20, 1960).

(24) Lloyd, R. D., "Preliminary Evaluation of Waspaloy Nickel-Base Alloy Sheet", Process Memorandum No. 14.096, The Marquardt Corporation, Van Nuys, California (March 19, 1959).

(25) Lund, C. H., "Physical Metallurgy of Nickel-Base Superalloys", DMIC Report No. 153, Battelle Memorial Institute, Columbus, Ohio (May 5, 1961).

(26) Mailer, R. R., "Compilation of Room Temperature and Short Time Elevated Temperature (1400 F) Tensile Properties of Precipitation Hardened René 41 Sheet", Report of the Republic Aviation Corporation, Farmingdale, New York (July 8, 1960).

(27) Marquardt Aircraft Company, Van Nuys, California, Preliminary Information on a United States Air Force Contract.

(28) Merriam, J. C., "Introduction to High Temperature Metals", Materials in Design Engineering, 51 (6), 143-158, Materials in Design Engineering Manual No. 172 (June, 1960).

(29) Mollica, R. J., "Very-Short-Time, High Temperature Properties of Titanium Alloy B120VCA, Magnesium Alloy ZE10A, Two PH Stainless Steels", Technical Memo MT-M57J, Chrysler Corporation, Detroit, Michigan (July 15, 1958).

(30) Moon, D. P., and Simmons, W. F., "Selected Short-Time Tensile and Creep Data Obtained Under Conditions of Rapid Heating", DMIC Report No. 130, Battelle Memorial Institute, Columbus, Ohio (June 17, 1960).

(31) National Research Corporation, Cambridge, Massachusetts, "Development of Tantalum Tungsten Alloys for High Performance Propulsion System Components", Preliminary information on a Naval Contract.

(32) National Research Corporation, Cambridge, Massachusetts, "Recent Experimental Results", Report on Contract NOrd 18787 (January 1, 1960).

(33) Northrop Aircraft, Incorporated, Hawthorne, California, "The Effects of Temperature Time-Stress Histories on the Mechanical Properties of Aircraft Structural Metallic Materials", Bimonthly Progress Report 5, NOR 59-247 on Contract AF 33(616)-5769 (January 15, 1959).

(34) Ogden, H. R., "Physical and Mechanical Properties of Tantalum", DMIC Memorandum No. 32, Battelle Memorial Institute, Columbus, Ohio (August 28, 1959).

(35) Pugh, J. F., "Relations Between the Elastic Moduli and Plastic Properties of Polycrystalline Pure Metals", Philosophical Magazine, 45, 823-843 (August, 1954).

(36) Roe, W. P., and Kattus, J. R., "Tensile Properties of Aircraft-Structural Metals at Various Rates of Loading After Rapid Heating", WADC-TR-58-199 (Part III), at Southern Research Institute, Birmingham, Alabama, on Contract AF 33(616)-424 (July, 1957).

(37) Schmidt, F. F., and Ogden, H. R., "The Engineering Properties of Columbium and Columbium Alloys", DMIC Report No. 188, Battelle Memorial Institute, Columbus, Ohio (September 6, 1963).

(38) Schmidt, F. F., and Ogden, H. R., "The Engineering Properties of Tantalum and Tantalum Alloys", DMIC Report No. 189, Battelle Memorial Institute, Columbus, Ohio (September 13, 1963).

(39) Schmidt, F. F., and Ogden, H. R., "The Engineering Properties of Molybdenum and Molybdenum Alloys", DMIC Report 190, Battelle Memorial Institute, Columbus, Ohio (September 20, 1963).

(40) Schmidt, F. F., and Ogden, H. R., "The Engineering Properties of Tungsten and Tungsten Alloys", DMIC Report No. 191, Battelle Memorial Institute, Columbus, Ohio (September 27, 1963).

(41) Semchyshen, M., and Barr, R. Q., "Investigation of the Effects of Hot-Cold Work on the Properties of Molybdenum", WADC-TR-56-454 of the Climax Molybdenum Company of Michigan, Detroit, Michigan, on Contract AF 33(616)-2861 (January, 1957). AD 110708.

(42) Semchyshen, M., and Barr, R. Q., "Mechanical Properties of Molybdenum and Mo-Base Alloy Sheet", Climax Molybdenum Company of Michigan, Detroit, Michigan, paper presented at the Third Pacific Area National Meeting of ASTM (October 15, 1959).

(43) Simmons, W. F., "High Temperature Materials, Their Strength Potential and Limitations", Proceedings of the Fourth Sagamore Ordnance Material Conference, August 21, 22, and 23, 1957.

(44) Southern Research Institute, Birmingham, Alabama, "Short-Time Elevated Temperature Mechanical Properties of Metals Under Various Conditions of Heat Treatment, Heating and Exposure Time and Loading", Summary Report for the Army Ballistics Missile Agency on Contract DA-01-009-ORD-494 (July 17, 1959).

(45) Sye, J. H., "Tensile and Stress-Rupture Properties of Waspaloy, René 41, and AF-1753 Bar Stock at 1700 F", Research Memo No. 59, of Universal-Cyclops Steel Corporation, Bridgeville, Pennsylvania (November 3, 1958).

(46) Tietz, T. E., Wilcox, B. A., et al., "Mechanical Properties and Oxidation Resistance of Certain Refractory Metals", Final Report of Stanford Research Institute, Menlo Park, California, on Contract NOas 58-366-d (January 30, 1959). AD 214829.

(47) Torti, M. L., "Development of Tantalum-Tungsten Alloys for High Performance Propulsion System Components", Quarterly Report No. 6 of the National Research Corporation, Cambridge, Massachusetts, on Contract NOrd 18787 (July 10 to October 9, 1960).

(48) Universal Cyclops Steel Corporation, Bridgeville, Pennsylvania, "Technical Data Sheets – Short Time Compressive Properties of Molybdenum and Mo-0.5 Ti Alloy Bar", Technical Bulletin No. 2.1.3 (October, 1960).

(49) Voorhees, H. R., and Freeman, J. W., "Report on the Elevated-Temperature Properties of Aluminum and Magnesium Alloys", ASTM Publication 291, Philadelphia, Pennsylvania (October, 1960).

APPENDIX E

SELECTED BIBLIOGRAPHY ON FORGING LUBRICANTS

APPENDIX E

SELECTED BIBLIOGRAPHY ON FORGING LUBRICANTS

(1) Alico, John, "Direct Extrusion of Magnesium Alloys", Light Metal Age, 11 (3, 4), 10-11, 22-23, 26 (April, 1953).

(2) Barnes, R. S., and Cafcas, T. H., "Laboratory Evaluation of Metal Forming Lubricants", Lubrication Engineer, 10 (3), 147-150 (May-June, 1954).

(3) Billington, P. H., "The Dry Lubrication of Moving Parts", Mechanical World, 133 (3411), 468 (1953).

(4) Bowden, F. P., "Recent Studies of Metallic Friction, Parts I and II", Engineer, 198, 886-889 (December 24, 1954); 902-905 (December 31, 1954).

(5) Brown, A. E., "Dry Film Lubrications Put on Like Enamel", Industrial Finishing, 30 (4), 50 (1954).

(6) Christensen, L. M., "Press-Extruded Forgings", Machine Design, 25 (7), 124-127 (July, 1953).

(7) "Colloidal Graphite and Metal Forming", Canadian Metals, 15 (12), 52-54 (December, 1952).

(8) Dietrich, R. L., and Ansel, G., "Calculation of Press Forging Pressure and Application to Magnesium Forgings", Transactions ASM, 38, 709 (1947).

(9) "Extrusion Goes Commercial", Steel, 132, 106-108 (April 27, 1953).

(10) Feigin, L. M., "Device for Studying Metal Friction at High Temperatures", Zavodskaya Laboratoriya, 24 (7), 869-870 (1958).

(11) Feldman, H. D., "Untersuchunger über den Kraft-und Arbeitshedarf beim Kaltflick pressen verschiedener Stahlsorten", Stahl u. Eisen, 73, 165 (1953).

(12) Genders, R., "The Extrusion Defect", Journal of the Institute of Metals, 26, 237 (1921).

(13) Holden, H. A., "A Review of Phosphate Coatings for Assisting Cold Extrusion", Sheet Metal Industries, 30, 502-512 (June, 1953).

(14) Hurst, A. L., "Aluminum Extrusion – From the Heavy Press Program", Product Engineering, 26 (8), 150-153 (August, 1953).

(15) Motherwell, G. W., "Production Experience With the 18,000 Ton Press", paper presented at Aircraft Industries Association Meeting on December 4, 1951.

(16) Nadai, A., "The Forces Required for Rolling Steel Strips in Tension", Journal of Applied Mechanics, ASME Transactions, 61, PA-54 (1936).

(17) Perry, H. W., "Extrusion With Glass Lubrication", Metal Industry, 82, 87-88 (January 30, 1953).

(18) Polakowski, N. H., "The Compression Test in Relation to Cold Rolling", Journal of the Iron and Steel Institute, 161, 250 (1949).

(19) Richards, G. W., "Large Forging-Presses", Aircraft Production, 15 (5), 154-161 (May, 1953).

(20) Sabroff, A. M., and Frost, P. D., "The Extrusion of Titanium", WADC-TR-54-555, Battelle Memorial Institute, Columbus, Ohio (December, 1954).

(21) Sachs, George, and Eisbein, W., "Power Requirements and Flow Processes of Extruded Rounds", Mitteilungen der Deutchen Materialprüfungsantalt, Special Volume, 16, 67-95 (1931).

(22) Salz, Leon, "What are the Differences in Wire Drawing Lubricants", Wire and Wire Products, 27 (11), 1180-1184 (November, 1952).

(23) Schor, Harold, "Metallurgical Aspects of Large Pressings, Forgings, and Extrusions for Aircraft", Metal Progress, 63 (3), 111-114, 182, 184, 186 (March, 1953).

(24) Schroeder, William, and Webster, D. A., "Press-Forging Thin Sections", Transactions ASME, 71, 289 (1949).

(25) Sejournet, J., "Complex Parts Produced by Hot Extrusion Process", Metals and Alloys, 31 (3), 56 (1950).

(26) Sheinhorz, I., McCullough, H. M., and Zambrow, J. L., "Method for Evaluation of Lubricants in Powder Metallurgy", Journal of Metals, 6 (5), 515 (May, 1954).

(27) Slater, I. G., "Extrusion of Aluminum", Metal Industry, 86, 464-468 (June 3, 1955).

(28) Snow, H. A., "The Influence of Soluble Oils on the Surface Finish in the Hot Rolling of Aluminum and its Alloys", Sheet Metal Industries, 31 (327), 601-607 (July, 1954).

(29) Sonntag, Alfred, "MoS_2 Simplifies Extreme Pressure Lubrication", Iron Age, 173 (5), 138-140 (May 20, 1954).

(30) Thompson, J. B., Turrel, G. C., and Sandt, B. W., "The Sliding Friction of Teflon", Plastics, 20 (215), 213-214 (June, 1955).

(31) Tracy, G. S., "Forging of Ti Requires Special Techniques", Machinery, 59 (10), 202-205 (June, 1953).

(32) "What's Ahead the Next Ten Years in Presswork and Forging", American Machinist, 96 (24), D-D11 (Mid-Nov., 1952).

UNCLASSIFIED

UNCLASSIFIED